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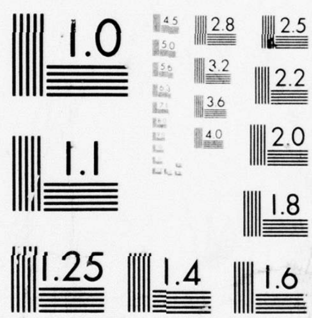

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Mechanism of Enhanced Toughness In Martensitic Alloys

Final Report

For the Period

October 1, 1975 - September 30, 1976

Contract No. N00019-76-C-0149

Prepared for

Department of the Navy
Naval Air Systems Command
Washington, D.C. 20361

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The objective of this program was to establish the role of heat treatment parameters on the fracture toughness of ultrahigh strength alloys. Microstructural examination and plane strain fracture toughness tests were conducted on alloys 4340 and silicon modified 4340. Heat treatments included both direct as well as step quench schedules. Solutioning temperatures ranged from 870°C to 1200°C, and step quenching temperatures ranged from 300°C to 1100°C.

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In Martensitic Alloys

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CONTENTS

I. ABSTRACT

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II. INTRODUCTION

The emergence of fracture mechanics and the concept of plane strain fracture toughness, K_{IC}, as a quantitative means of a material's tendency to fail in a brittle manner is playing an increasingly important role in the selection and use of high strength alloys[(1)]. Such steels are often chosen according to this relative fracture toughness at different strength levels, and the use of these materials is limited by their low fracture toughness at high strength levels. Hence there is a great need to develop an understanding of the mechanisms involved in the fracture of these alloys, and an understanding of the variables which control the fracture toughness. It is the recognition and utilization of these factors which will provide the basis for the design of new high strength alloys with superior properties.

Maraging steels, as a class of alloys, exhibit one of the best combinations of strength and toughness available, better than conventionally treated low alloy steels such as 4140 and 4340[(2)]. However, cost limits their use except where absolutely necessary. Results have indicated that the long associated poor toughness of these very high strength low alloy steels can be significantly improved, approaching the values obtained for the maraging steels, without the high cost. This has been accomplished by altering only the heat treatment procedures. Furthermore the fracture toughness levels have been achieved without a reduction in strength.[(3-6)]

In order to understand the role of microstructure on the fracture toughness it is necessary to characterize the microstructure carefully. This investigation has utilized optical, transmission, and scanning electron microscopy in conjunction with mechanical testing.

The main objective of this program has been to establish the critical role of heat treatment parameters on the plane strain fracture toughness of alloys 4340 and silicon modified 4340.

III. EXPERIMENTAL PROCEDURE

A. Material

The program examined two alloys, aircraft quality 4340, mil specification AMS 6359, and a silicon modified 4340. The silicon modified 4340 or 300M, as it will be referred to in this report, was obtained from the Boeing Company in Seattle, Washington. The Boeing Company's specification is BMS 7-26. This alloy contained 1.6 wt. percent silicon in addition to the normal 4340 composition. The alloy was vacuum remelted.

B. Heat Treatment

A controlled atmosphere tube furnace was utilized for all high temperature austenitization treatments. This furnace maintained a temperature within ± 5°C. Test specimens were either quenched directly from the solution temperature into an agitated oil bath or step quenched into a salt bath prior to oil quenching. All tempering was carried out for an hour. Standard step quench involved holding for 30 minutes before oil quenching. In order to evaluate the effect of holding time, one series of tests was carried out using several holding times varying from 10 seconds to 2 hrs. These specimens were stepped into salt baths to insure rapid thermal equilibrium for the short hold times. All tests were carried out on standard ASTM compact tension fracture toughness specimens. Sections for optical and transmission electron microscopy were taken from mid-thickness areas adjacent to the fracture path.

For the isothermal transformation to bainite and retained austenite, 4340 and 300M were again used. By varying the isothermal holding time the amount of retained austenite was controlled. To insure a fine lower bainite structure the alloy was transformed in salt baths just above the M_s temperature. Retained austenite has been qualitatively estimated from electron microscopy.

C. Fracture Toughness Tests

All fracture toughness tests were carried out using ASTM specified compact tension test specimens. Testing was carried out at room temperature. Specimens were 5/8" thick and were tested in the longitudinal direction. All specimens were machined to final dimensions prior to heat treating except for the starter slot. This was added after heat treating by using a .008" thick grinding wheel. A 22,000 lb Instron Lawrence dynamic test system was used for all testing, including fatigue precracking. Stress intensities were calculated according to ASTM standards E399.

D. Microstructural Analysis

Sections for optical and transmission electron microscopy were taken from the midsection of the K_{Ic} specimens. For optical metallography polished specimens were etched in a combination picral and nital etch. Thin foil preparation was carried out using both the window technique

and the jet polishing technique, (Fischione unit). Two electrolytes, glacial acetic acid + perchloric acid and glacial acetic acid + chromium trioxide,were used. The best results were obtained from the latter in conjunction with the window technique. The exact composition of the electrolyte and the polishing conditions are given below:

Electrolyte:

Glacial Acetic Acid	135 ml
Chromium Trioxide	25 gms
Water	7 ml

Polishing Conditions:

Temperature	10-15°C
Voltage	25 volts
Current density	0.1-0.2 Amp/cm^2

The starting material was obtained in 10-15 mil thickness sections by cutting heat treated specimens with a 1/32" abrasive wheel. Sections were cut while flooded with water. A very low cutting rate was employed. These 10-15 mil sections were then carefully ground to about 5 mils thickness. From this thickness final polishing by either the window or the jet polishing technique was carried out.

E. Fracture Morphology

Secondary electron scanning microscopy was carried out at 25 KV. For each specimen the region adjacent to the fatigue precrack was examined, since this is the region of crack initiation during the K_{Ic} test. This region typically extends about .030" from the fatigue crack.

IV. EXPERIMENTAL RESULTS

A. Fracture Toughness

Tables 1-4 and Figures 1-11 summarize the plane strain fracture toughness results obtained during this program. All the data were valid except those marked by an asterisk, where P_{max}/P_q exceeded 1.10. Figures 1-9 graphically illustrate the effects of various heat treatments. For both alloys the conventional heat treatment consisted of solutioning at 870°C (1600°F) for one hour followed by oil quenching. The effect of increasing the austenitizing temperature from 870 to 1200°C is apparent in Figure 1 for the as quenched test condition. Increasing the temperature from 870 to 1100°C resulted in only a small increase from about 35 ksi-in$^{1/2}$ to 43 ksi-in$^{1/2}$. Not shown in this figure, but included in the tables is data for a 1150°C treatment. This heat treatment resulted in a toughness of about 52 ksi-in$^{1/2}$. Further increases in the solution temperature to 1200°C and direct quenching produced the highest toughness in the as quenched condition, 66 ksi-in$^{1/2}$.

Aside from the increased toughness due to raising the solution temperature, Figure 1 also reveals the effect of step quenching to 870°C for successively longer holding times. Stepping to 870°C for 30 seconds did not reduce the toughness, but it is doubtful that 30 seconds in a salt bath was sufficient to reduce the specimen temperature to 870°C. Hence, little change was expected. Continuing work during the 2nd year will include

instrumented cooling and quenching data to confirm actual equilibrium time-temperature relationships for step quenching. Five minutes is sufficient, however, to reduce the specimen to 870°C. Figure 1 demonstrates the negative effect on the fracture toughness. Holding for 30 minutes and one hour reduced the toughness even further. Not shown in the figures, but included in the tables are the results of short time 15 minute holds at 1200°C. The results demonstrated that holding for 1 hour was not necessary.

The trends present in Figure 1 demonstrate the major effects of heat treatment which were true for both alloys and for all tempering temperatures up to 175-200°C. These effects can be summarized as follows:

(1) Direct quenching from 1200°C produced the highest toughness.

(2) Direct quenching from lower temperatures produced successively lower toughnesses.

(3) Step quenching to 870°C reduces the toughness with the extent of the toughness reduction dependent on the holding time at 870°C.

The effects of tempering on 4340 are shown in Figure 2. There are several important features on these curves. First, for low tempering temperatures, below 175°C, the 1200°C clearly has superior toughness while the 1100°C treatment provides little benefit over that of the 870°C treatment.

Second, for these tests, the conventional 870°C treatment produced a continuously increasing toughness as the tempering temperature increases, while the higher temperature treatments show a definite embrittlement reaction at about 200°C.

Step quenching to just above the M_s temperature for either partial or complete transformation to bainite resulted in mixed results. Figure 3 indicates that tempering after transformation improved the toughness and that this trend is also shown by specimens partially transformed to bainite. Hence, even for the bainitic structure a large grain size was necessary. Also shown in Figure 3 are a 1200°C → 315°C for 10 second step quench and a 1200°C → 1100°C step quench for 1 minute. Both these resulted in very high toughness levels. However, it is doubtful that either specimen reached the step temperature before final oil quenching. Hence, these structures should be martensitic. Again instrumented quenching data is necessary.

A comparison of Figures 2 and 3 is shown in Figure 4. Clearly the direct quench or very rapid step quench produced the best toughness if tempering temperatures below 175°C are utilized.

The fracture toughness results obtained by step quenching to just above the M_s temperature and holding for various times are summarized in tables 3 and 4 for both 4340 and 300M. For alloy 4340 the results show that there were several combinations which yielded plane strain fracture

toughness data above 90,000 psi-in$^{1/2}$. Table 3 also reveals that the 1200°C → 1100°C can result in very high toughness, if the holding time at 1100 is short. Prolonged holding produced a sharp drop in toughness.

Alloy 300M behaved in basically the same manner as 4340. Again Figure 5 indicates that the direct quench from 1200°C produced the highest toughness independent of the tempering temperature employed. One difference, however, is the lack of the severe temper embrittlement reaction present in alloy 4340. For alloy 300M the direct quench from 1200°C was superior regardless of the tempering temperatures used in this study to date.

Step quenching to either partial or complete transformation to bainite produced results comparable to alloy 4340. Again, prior treatment at 1200°C was necessary to achieve maximum toughness. A comparison of Figures 5 and 6 shows that the direct quench from 1200°C produced the highest toughness, although the difference was small. However, there was not sufficient material to test the effect of higher austenitizing temperatures. The effect of a small amount of transformation to bainite is not significant for alloy 300M, whereas it produced a significant drop in the toughness for alloy 4340.

Figures 8 and 9 compare the effects of tempering on the toughness of 4340 and 300M for both the 870°C and the direct quench from 1200°C heat treatment. For the 870°C treatment the curves are very similar. The difference in toughness at tempering temperatures above 150°C was due to

the tempering resistance of 300M. The added silicon content has a well known effect on the tempering reaction. For the 1200°C treatment the maximum toughness occurred at a tempering temperature of 175°C for both alloys. However, the strength of alloy 300M is about 20,000 psi greater than 4340 at this tempering temperature.

In order to compare this data to other published data, these maximum toughness levels have been superimposed on data for these alloys and other ultrahigh strength steels, Figures 10-11. Clearly the modified treatments resulted in a fracture toughness equivalent to the 18% Nickel maraging steels at equivalent strength levels.

B. Microstructure Results

The microstructure of each alloy was examined by both optical and electron microscopy. Both replica and foil techniques were utilized. Grain size calculations were carried out for various austenitizing treatments. The results are summarized as follows:

Solution Treatment	Grain Size, ASTM	
	4340	300M
as received	9	6.5
870°C/1 hr.	9	6.5
1000°C/1 hr	9	4
1100°C/1 hr.	5.5	2
1100°C/2 hr.	-	2
1150°C/1 hr.	3	2
1200°C/15 min.	1	-
1200°C/30 min.	1	-
1200°C/1 hr.	1	2
1200°C/1 hr – 1100°C/15 min	-	2
1200°C/1 hr – 1100°C/1 hr	-	2

Clearly 4340, a rolled plate material had a finer grain size initially. However, after solutioning at 1200°C, both alloys have ASTM grain size of about 1-2. Step quenching does not result in a further increase in grain size. Since the grain size is very large after the 1100°C/1 hr. treatment but the fracture toughness is unchanged, and since step quenching did not result in as high a fracture toughness as the direct quench, then grain size per se can not be controlling the fracture toughness of these alloys.

Obviously, the heat treatment, fracture toughness, and microstructure are all interrelated and thus it is difficult to discuss any one separately. The heat treatment schedules and the alloys themselves are complex enough in that it is difficult to give a simplified explanation for any microstructural interpretation let alone fracture toughness interpretation. However, several microstructural observations can be made which all effect the final result, the fracture toughness. Figures 18-28 contain optical and electron micrographs including replicas and thin foil electron micrographs.

Figure 18 reveals the banded structure of alloy 4340 in the as-quenched condition and after solutioning at 870°C for 1 hour. While the banding which is due to segregation is no longer visible , the segregation probably remains. Electron microprobe analysis is required to confirm this. Figure 21 which shows as-received 300M does not reveal banding. However, alloy 300M was obtained as a forging whereas alloy 4340 was supplied under AMS 6359 specifications for plate material.

Large dark etching plates are visible in Figure 18. These are most probably martensitic plates which have been autotempered during the quench. In some of these dark etching plates a parallel series of sub-units are visible, when viewed under very high magnifications, i.e. 2000X. Unfortunately, it is impossible to uniquely identify these optically, and the probability of finding an isolated area with thin foil is low.

Examples of 4340 given a bainitic heat treatment are shown in Figures 19 and 20. The effect of an initially banded segregated structure on the subsequently heat treated structure is evident in Figure 19. Referring to the TTT curves it is seen that this sample should contain a small amount of bainite. The segregated nature of the bainite in a matrix of martensite is clearly visible. Upon cooling to room temperature the remaining austenite underwent transformation to martensite and retained austenite. A one hour hold at 315°C resulted in essentially complete transformation to bainite.

Figure 20 reveals the microstructure of 4340 after an initial austenitization of 1200°C. The large grain size makes the bainitic structure easier to identify. The 3 min. hold at 335°C results in about 25% transformation to bainite with the remainder essentially transforming to martensite upon quenching to room temperature. The carbide precipitation can be easily distinguished in higher magnification replicas which also clearly show the duplex bainitic-tempered martensitic structure, as opposed to the

essentially fully bainitic structure present in specimens held for 1 hr. in salt prior to quenching.

Microstructures of alloy 300M corresponding to various heat treatments for which fracture toughness data was also obtained are included in Figures 21-23. There are several noteworthy differences between 4340 and 300M. First, in Figure 21,very fine residual carbides are visible as small black spots about 0.5 mm in diameter. There were no residual carbides in alloy 4340 for this heat treatment. The high austenitizi treatment did produce a network of dark etching plates surrounding most prior austenite grain boundaries, Figure 21. These are present in the fine grain case, but not very evident at low magnification due to the fine grain si

Isothermally transformed structures are shown in Figure 2 For alloy 300M,a 3 min. hold at 315°C or 350°C should only produce about 2-5% bainite for a 1-1/2% silicon content. The remaining austenite has evidently transformed to martensite for the case of an initial 870°C solution treatment. Retained austenite is found, however, within the bainitic ferrite laths as shown in thin foil transmission electron micrographs.

By holding for 1 hour at 310°C about 75% transformation to bainite occurs according to the TTT curve after an initial 900°C solution. This is in good agreement with Figures 23. Figure 23a has large white area

representing either retained austenite or untempered martensite formed during the quench to room temperature. Tempering at 200°C transformed these white regions into a dark etching structure, Figure 23b. Similar micrographs for a prior 1200°C solution treatment are included in Figures 23c and 23d for 300M. However, there is one important difference evident in Figure 23d. Again after holding for 1 hour the structure was essentially bainitic with a small amount of white islands. However tempering for 1 hour at 200°C did not eliminate these white islands. If these were untempered martensite they would have been tempered. Hence, the stability of these regions is different and depends on the solutioning temperature. It is impossibl to establish the structure of these regions by optical microscopy. Transmission electron microscopy is underway as part of the second year program in order to determine the effect of solutioning temperature, silicon content and tempering temperature on the structures present under these conditions.

Transmission electron microscopy results are shown in Figures 24 and 25 for alloy 4340 for both large and fine grain size for both direct and step quench situations. Thin foil analysis for two step quench treatments to 1100°C are included, Figures 24 c-f and Figure 25. Of possible significance is the fact that the sample held for 30 minutes was extensively twinned while the specimen held at 1100°C for only 10 minutes contained very few twins. Referring back to Table 3, it is clear that a significant drop in fracture toughness occurred. The 10 minute hold had a toughness of 93,460 psi-in$^{1/2}$ while the 30 minute hold had only 68,000 psi-in$^{1/2}$ fracture toughness. Also note the large variation in plate sizes for this and

other sections as well. Finally, extensive carbide precipitation is visible throughout all the tempered structures.

Figures 26-28 reveal the structure of 300M. Bright field, dark field and corresponding diffraction patterns are shown. The retained austenite distribution can be seen in the figures. In Figure 27 a and b, the dark films between the ferrite laths are austenite. However, Figure 27c reveals that the structure is not entirely bainitic, i.e. martensitic regions are visible. This shows that some austenite was not stabilized after 1 hr. at 350°C given a prior 870°C solution treatment. Figure 28 is also 300M isothermally transformed to bainite. However, the transformation temperature was 310°C instead of 350°C. This resulted in a finer structure as expected. Again, retained austenite was present but generally on a much finer scale. Note that there are no carbides in this bainite, only ferrite and films of retained austenite. This form of bainite has been reported in the literature.

C. Fracture Morphology

The fracture morphology of both alloy 4340 and 300M were extensively examined. Mixed structures were present for nearly all heat treatment conditions. Figure 29a clearly shows the mixed ductile and cleavage modes present. Tempering reduces the amount of cleavage, Figure 29 b and c. The interface between the fatigue precrack and the initial crack growth region is visible in Figure 29c. In all cases only regions near

the fatigue crack were examined. The 1200°C/oil quench fracture morpholog is shown in Figure 30a. The high magnification micrograph indicates shallow d type fracture indicating very localized plastic flow. Again both dimpled and quasicleavage areas were present. Step quenching to 870°C resulted in a somewhat more intergranular-fracture mode, Figure 30b.

Step quenching from 870°C to 315°C to form a duplex structure resulted in two distinct fracture modes, Figure 31a. However, additional tempering at 200°C eliminated the intergranular cleavage and also increased the fracture toughness somewhat. Recall, however, that the highest toughness was achieved by the direct quench from 1200°C, or by step quenching for very short times prior to oil quenching. Micrographs for the 1200 → 335°C step quench to form a duplex martensitic-bainitic structure are included in Figure 32. The transition from a ductile to a brittle fracture is obvious as the holding time at 335°C is increased from 1 minute to 1 hr.

It is difficult in some cases to identify morphology features which can be uniquely associated with either a high or low toughness level. Figure 33 contains micrographs from K_{Ic} specimens which were stepped from 1200°C – 1100°C for 1 min (Fig. 33 a and b) and 1 hour (Fig. 33c). All three figures are similar in that they exhibit coarse intergranular features which reveal localized plastic flow along grain boundaries. However referring

to Table 3, it is seen that the toughness of 33 a & b is
the toughness of 33c is very low.

Alloy 300M did not exhibit intergranula
the large grained specimens, Figure 35. However, ter
produced a much more ductile fracture surface compar
surface of as quenched test specimen. For this case in
can be associated with improved toughness. However,
Figure 34b and 35b demonstrates that it is very
the relative toughness of specimens by just examining th
For this case both specimens exhibit a ductile failure n

D. Hardness Testing

A limited amount of hardness testing ha
preliminary to tensile testing. This data is shown in Fi
4340 and 300M respectively. For the direct quench fro
1200°C there was little change in hardness for either al
In each case the samples transformed for 60 minutes ha
due to the formation of a large amount of bainite rather
The choice of transformation temperatures was based o
peratures and was chosen in order to transform about 2
However, judging from the measured hardnesses the M
be lower than anticipated. Bainitic hardness levels app
be obtainable. A series of transformation temperatures

during the second year program in order to establish the

that can be achieved by this method.

V. DISCUSSION

The results of this program have demonstrated that very high f toughness levels can be achieved. In fact, by direct quenching from 1 rather than step quenching, even higher levels of toughness can be ach while maintaining very high strength levels. A unique explanation for observed property changes is difficult to make based on microstructur analysis. Both the fine grain and the large grain size are martensitic direct quenching. Both contain lath and plate martensite. There are variations in the amount of twinning in some cases, but any firm concl must wait until additional thin foils are examined for variations in the and morphology of the twins. Variations in the amount of retained aus have been observed, but it is difficult to positively conclude that this a for the improvement. The most informative results are the trends in ture toughness during step quenching, and the fact that 1200 and not 11 1100°C is necessary for maximum toughness. Electron microscopy w continue during the 2nd year effort on these samples in order to identif variations in austenite and twin density. In addition to austenite and tw considerations, other microstructural possibilities are lath size and p size both of which might control a critical fracture path length. Again large number of foils must be prepared and examined before conclusio be reached. Electron microscopy will continue throughout the 2nd yea in order to obtain sufficient data to make a meaningful evaluation.

There is one additional potential explanation which the data supports. Segregation of both the alloy elements such as nickel as well as trace level impurity elements may cause significant changes in the fracture toughness. The reduced toughness levels produced by step quenching support this theory, especially since increasing the holding time at either 1100°C or 870°C severely reduces the fracture toughness. It is important to note that at these intermediate step temperatures ranging from 870 to 1100°C, no precipitation, transformations, or grain growth occur. Thus in addition to variations in austenite content and morphology as well as variations in the substructure of the martensite, variations in segregation due to both grain size and step quench temperature and time are to be taken into account. Auger electron spectroscopy will be carried out during the second year in order to quantitatively establish the existence of segregation as a possible mechanism controlling the toughness of these alloys.

High temperature solution treatments alter the shape of the TTT curve for 4340 and silicon modified 4340. Figures 14-16 relate to the TTT curves for each alloy as a function of austenitizing temperature. A comparison of Figures 14 and 15 demonstrates the effect of the austenitizing temperature. Note that the M_s temperature increases about 25°C, and that the nose of the curve is retarded. This increases the upper bainitic hardenability. However, the kinetics of the lower bainitic transformation are essentially unchanged. Figure 16 is for a silicon modified 4340 with 1% silicon. The silicon content also affects the kinetics of the bainite reaction. For example, 4340 transformed

at 300°C requires 5 minutes to reach 70% transformation to bainite whereas with 1% silicon, one hour is required to produce 70% transformation. Figure 17 indicates the amount of austenite retained at any transformation temperature after the reaction has stabilized after a prior 900°C solution treatment.

The fracture toughness of each alloy isothermally transformed to bainite was not as high as the fracture toughness of the martensitic structures. This may be due to inherently lower toughness in the bainitic structure or it could relate to a segregation model again. Many studies have shown that as the carbon content is raised, the optimum structure for maximizing the toughness, goes from martensitic to bainitic. In general, the transition carbon level is about .4% by weight. Hence, both alloys are in the transition region. If, however, the austenite is stabilized by a supersaturation of carbon within the austenite, then the overall carbon content of the martensite may be lowered. This would be expected to improve the toughness of the martensite. However, one might also expect the strength to drop just on the basis of lower strength martensite.

The advantage of a short time step quench to just above the M_s temperature would be to reduce the thermal gradient prior to quenching through the martensite transformation range. In order to do this effectively, it is necessary to know how long it takes test specimens, or actual components, to reach the temperature of the bath. Work during the second year program will include instrumented quenching data in order to optimize the actual step quench time for several temperatures.

VI. SUMMARY & CONCLUSIONS

It appears for the work to date in this program that martensitic structures have better toughness than bainitic structures, even though bainitic structures tend to have more retained austenite. However, the bainitic structures produced to date have been fairly coarse and a further reduction in the isothermal transformation will be utilized in order to refine the structure and increase the strength of the bainitic structure.

The effects of step quenching over a range of temperatures for varied holding times have been demonstrated. A complete description of these effects will require additional testing as outlined in the second year program. A better understanding of the stability of the austenite under different conditions has been obtained. The stability of the retained austenite depends on both the amount of silicon present and the austenitizing temperature and quenching conditions. Additional work utilizing electron microscopy, X-ray diffraction and auger electron spectroscopy is required. This work will form part of the second year program. The primary results are the variations in the fracture toughness levels achieved by different heat treatments. These are summarized as follows:

(1) For maximum toughness initial austenitization at 1200°C, was required.

(2) Maximum toughness was obtained by using several different heat treatment schedules. In addition to the direct quench from 1200°C, step quench treatments to 1100, 870°C or 315°C all produced high toughness levels if the holding time was short.

(3) The 1200°C/1 hr → 1100°C intermediate heat treatment produced interesting results relative to the importance of the 1200°C initial temperature. Note that upon holding for successively longer times at 1100°C, the toughness dropped from 91, 370 psi-in$^{1/2}$ for a 1 minute hold to 68,000 for a 30 minute hold for a 175°C temper.

(4) For specimens held one hour at 315 or 335°C, the toughness, although severely reduced, is essentiallv independent of the prior austenitizing temperature.

(5) Valid plane strain fracture toughness results of over 90,000 psi-in$^{1/2}$ were achieved for alloy 4340 combined with tensile strengths of 300,000 psi. K_{Ic} values of over 85,000 psi-in$^{1/2}$ have been obtained for alloy 300M at equivalent strength levels.

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Table 1

Fracture Toughness Data for Alloy 4340

I.D.	Heat Treatment: Austenitizing Temp. °C/hr.	Quench	Tempering Temp. °C/hr.	K_{IC} psi-in$^{1/2}$	K_{max} psi-in$^{1/2}$	P_m/P_Q
AC2	870/1	oil	AQ	38,560	38,560	1.00
AC14	870/1	oil	AQ	34,390	34,390	1.00
AC18	870/1	oil	AQ	33,510	37,720	1.02
A19	870/1	oil	125/1 hr	42,240	42,240	1.00
AC3	870/1	oil	200/1 hr	61,460	67,920	1.04
AC5	870/1	oil	200/1 hr	64,600	64,600	1.00
AC6	870/1	oil	280/1 hr	77,470	88,180	1.03
AC7	870/1	oil	280/1 hr	75,120	80,560	1.03
AC8	870/1	oil	350/1 hr	92,630	113,33	1.12*
AC9			350/1 hr	103,290	117,810	1.05
AC12	1100/1	oil	AQ	42,970	42,970	1.00
AC13	1100/1	oil	AQ	43,130	43,130	1.00
AC53	1100/1	oil	125/1	45,620	50,760	1.07
AC27	1100/1	oil	175/1	58,170	61,960	1.01
AC26	1100/1	oil	175/1	61,110	63,670	1.01
AC28	1100/1	oil	200/1	61,440	61,440	1.00
AC15	1200/1	oil	AQ	67,300	79,560	1.11
AC22	1200/1	oil	AQ	46,780	62,870	1.18*
AC30	1200/1	oil	AQ	65,880	76,400	1.07
AC24	1200/1	oil	125/1	83,074	89,660	1.04
AC25	1200/1	oil	125/1	83,990	83,990	1.00
AC23	1200/1	oil	225/1	66,360	85,090	1.17*

Table 1 (cont'd)

Fracture Toughness Data for Alloy 4340

I.D.	Heat Treatment: Austenitizing Temp. °C/hr.	Quench	Tempering Temp. °C/hr.	K_{IC} psi-in$^{1/2}$	K_{max} psi-in$^{1/2}$	P_m/P_Q
AC29	1200/1	oil	175/1	85,350	98,390	1.08
AC31	1200/1	oil	280/1	59,580	70,070	1.08
AC54	1200/1	oil	350/1	72,610	91,520	1.11*
AC55	1200/1	oil		80,790	81,900	1.01
AC47	1200/1→870/½hr	oil	AQ	51,660	53,580	1.01
AC51	1200/1→870/½hr	oil	AQ	52,970	63,410	1.10
AC48	1200/1→870/½hr	oil	175/1	65,540	70,680	1.01
AC49	1200/1→870/½hr	oil	175/1	60,960	72,410	1.12*
AC37	1200/1→870/½hr	oil	200/1	60,720	82,830	1.13*
AC38	1200/1→870/½hr	oil	200/1	60,970	70,640	1.08*
AC41	1200/1→870/½hr	oil	280/1	52,250	58,080	1.06
AC42	1200/1→870/½hr	oil	280/1	52,970	59,120	1.06
AC44	1200/1→870/½ hr	oil	350/1	85,640	96,060	1.01
AC45	1200/1→870/½hr	oil	350/1	84,730	96,300	1.06
AC43	1200/1→870/½hr	oil	150/1	59,370	68,770	1.06
AC46	1200/1→870/1 hr	oil	AQ	52,890	55,960	1.03
AC50	1200/1→870/5 min	oil	AQ			
AC52	1200/1→870/5 min	oil	AQ	56,660	67,210	1.11*
AC39	1200/1→870/30 sec	oil	AQ	62,620	75,380	1.11*
AC40	1200/1→870/30 sec	oil	AQ	64,960	74,210	1.10

Table 1 (cont'd)

Fracture Toughness Data for Alloy 4340

I.D.	Austenitizing Temp. °C/hr.	Quench	Tempering Temp. °C/hr.	K_{IC} psi-in$^{1/2}$	K_{max} psi-in$^{1/2}$	P_m/P_Q
AC56	870/1→315/3 min	oil	AQ	40,340	48,070	1.10
AC57		oil		39,430	45,510	1.07
AC58			200/1	51,400	54,540	1.03
AC59				52,960	58,980	1.03
AC60	870/1→315/60 min	oil	AQ	49,540	59,630	1.04
AC61				52,480	63,100	1.06
AC62			200/1	57,160	66,830	1.09
AC63				58,590	62,430	1.02
- - -	- - -	- - -	- - -	- - -	- - -	- - -
AC82	1200/1→300/3 min	oil	200/1	79,240	88,800	1.05
AC83				80,400	84,730	1.03
AC80	1200/1→300/60 min	oil	200/1	62,530	62,530	1.00
AC81				65,370	65,370	1.00
AC87	1200/1→315/10 sec	oil	175/1	90,220	100,830	1.04
AC84			200/1	88,420	93,410	1.02
AC85				88,150	99,880	1.08
AC72	1200/1→335/1 min	oil	200/1	89,690	94,590	1.02
AC64	1200/1→335/3 min	oil	AQ	58,150	66,280	1.06
AC65				60,480	66,830	1.03
AC73			175/1	80,220	82,110	1.01
AC70			200/1	74,460	82,730	1.03
AC71				78,240	79,670	1.00
AC74			280/1	68,490	81,430	1.06
AC 75	1200/1→335/5 min	oil	200/1	82,670	88,670	1.01

Table 1 (cont'd)

Fracture Toughness Data for Alloy 4340

I.D.	Austenitizing Temp. °C/hr.	Quench	Tempering Temp. °C/hr.	K_{IC} psi-in$^{1/2}$	K_{max} psi-in$^{1/2}$	P_m/P_Q
AC66	1200/1→335/60 min	oil	AQ	43,490	50,720	1.08
AC67				45,890	45,890	1.00
AC68			200/1	53,930	58,050	1.01
AC69				47,160	57,390	1.10
AC79				50,530	58,450	1.04
AC86	1200/1→1100/1 min	oil	175/1	91,370	101,040	1.05
AC88			200/1	85,710	100,020	1.06
AC89	1200/1→1100/10 min	oil	175/1	93,460	107,610	1.05
AC34*			200/1	67,880	77,430	1.05
AC32	1200/1→1100/30 min	oil	200/1	72,050	76,510	1.02
AC33			200/1	63,040	69,090	1.03
AC76	1150/1	oil	AQ	50,690	53,130	1.01
AC77				53,710	57,300	1.04
AC78			175/1	70,070	76,410	1.04

*Possible variation in holding time at 1100°C. Additional tests to be carried out.

Table 2

Fracture Toughness Data for Alloy 300M

I.D.	**Heat Treatment** Austenitizing Temp. °C/hr.	Quench	Tempering Temp. °C/hr.	K_{IC} psi-in$^{1/2}$	K_{max} psi-in$^{1/2}$	P_m/P_Q
BC7	870/1	oil	AQ	36,520	54,090	1.13*
BC8	870/1	oil	AQ	36,480	53,550	1.11*
BC39	870/1	oil	175/1	68,450	80,865	1.01
BC10	870/1	oil	200/1	67,980	78,930	1.02
BC28	870/1	oil	200/1	65,490	76,860	1.02
BC11	870/1	oil	280/1	63,850	75,380	1.03
BC29	870/1	oil	280/1	66,600	73,850	1.02
BC30	870/1	oil	350/1	67,950	75,260	1.03
BC38	870/1	oil	125/1	45,940	62,410	1.07
BC27	870/1	oil	150/1	44,640	58,150	1.05
BC14	1100/1	oil	AQ	39,790	60,610	1.18*
BC13	1100/1	oil	AQ	42,820	61,760	1.13*
BC45	1100/1	oil	175/1	76,500	93,050	1.05
BC31	1100/1	oil	200	76,300	93,460	1.05
BC32	1100/1	oil	280	72,260	95,330	1.06
BC33	1100/1	oil	350	72,570	87,930	1.08
BC15	1200/1	oil	AQ	43,670	55,330	1.02
BC16	1200/1	oil	AQ	39,720	65,270	1.22*
BC37	1200/1	oil	125/1	53,010	75,200	1.07
BC48	1200/1	oil	125/1	53,350	76,320	1.08
BC49	1200/1	oil	150/1	59,480	81,310	1.08
BC40	1200/1	oil	175/1	89,540	107,390	1.05
BC46	1200/1	oil	175/1	77,670	106,060	1.13*

Table 2 (cont'd)

Fracture Toughness Data for Alloy 300M

I.D.	Heat Treatment: Austenitizing Temp. °C/hr.	Quench	Tempering Temp. °C/hr.	K_{IC} psi-in$^{1/2}$
BC17	1200/1	oil	200/1	80,140
BC18	1200/1	oil	200/1	85,530
BC35	1200/1	oil	200/1	82,810
BC19	1200/1	oil	280/1	78,080
BC36	1200/1	oil	280/1	83,630
BC20	1200/1	oil	350/1	73,930
BC41	1200/½hr	oil	200/5	79,370
BC42	1200/½hr	oil	200/5	79,120
BC43	1200/¼hr	oil	200/5	89,000
BC44	1200/¼hr	oil	200/5	85,720
BC47	1200/¼hr	oil	200/1	88,970
BC21	1200/1→870/½hr	oil	AQ	42,560
BC22	1200/1→870/½hr	oil	AQ	41,680
BC23	1200/1→870/½hr	oil	200/1	75,670
BC24	1200/1 870/½hr	oil	200/1	79,010
BC25	1200/1→870/½hr	oil	280/1	72,440
BC26	1200/1→870/½hr	oil	350/1	66,000

Table 2 (cont'd)

Fracture Toughness Data for Alloy 300M

I.D.	Austenitizing Temp. °C/hr.	Quench	Tempering Temp. °C/hr	K_{IC} psi-in
BC52	870/1 315/3 min	oil	AQ	
BC53			200/1	67,270
BC50	870/1 315/60 min	oil	AQ	43,200
BC51			200/1	60,730
BC56	1200/1 315/3 min	oil	AQ	49,570
BC34			200/1	84,050
BC54	1200/1 315/60 min	oil	AQ	46,730
BC55			200/1	61,630

*invalid test result.

TABLE 3 Fracture Toughness versus Heat Treatment for Alloy 4340

Austenitizing Temperature °C (for 1 hour)	Intermediate Holding Temperature °C	Final Tempering Temperature °C (for 1 hour)	Plane Strain Fracture Toughness, K_{IC} (Average Values)						
			Hold Time at Intermediate Temperature						
			10 sec.	1 min.	3 min.	5 min.	10 min.	30 min.	60 min
1200	300	As quenched							
		200			79,820				
	315	175	90,200						
		200	88,280						
	335	As quenched			59,320				44,690
		175			80,220				
		200		89,690	76,350	82,460			50,540
		280			68,490				
	1100	175		91,370			93,460	68,000	
		200		85,710			67,880*		67,000
870	315	As quenched			39,890				51,010
		200			52,180				57,880

*Possible error in heat treatment process, additional samples to be tested.

TABLE 4 Fracture Toughness versus Heat Treatment for Alloy 300M

Austenitizing Temperature °C (for 1 hour)	Intermediate Holding Temperature °C	Final Tempering Temperature °C (for 1 hour)	Plane Strain Fracture Toughness, K_{IC} (Average Values)						
			Hold Time at Intermediate Temperature						
			10 sec.	1 min.	3 min.	5 min.	10 min.	30 min.	60 min
870	315	As quenched							43,200
		200			67,270				60,730
1200	315	As quenched			49,570				47,730
		200			84,050				61,650

FIGURE

FIGURE

18	4340	(a) As received
		(b) 870°C/1 hr and oil quenched
		(c) 1200°C/1 hr and oil quenched
		(d) 1200°C/1 hr – 870°C/1/2 hr and oil quenched
19	4340	(a-c) 870°C/1 hr – 315°C/3 min and oil quenched
20	4340	1200°C/1 hr – 335°C/3 min and oil quenched
		(a) As quenched
		(b) Tempered at 200°C/1 hr
21	300M	(a) As received
		(b,c) 870°C/1 hr and oil quenched
		(d) 1200°C/1 hr and oil quenched
22	300M	(a) 870°C/1 hr – 315°C/3 min and oil quenched
		(b) 1200°C/1 hr – 315°C/3 min and oil quenched
23	300M	870°C/1 hr – 310°C/1 hr and oil quenched
		(a) As quenched
		(b) Tempered at 200°C/1 hr
		1200°C/1 hr – 315°C/1 hr and oil quenched
		(c) As quenched
		(d) Tempered at 200°C/1 hr.
24	4340	Bright field TEM micrographs
		(a,b) 870°C/1 hr and oil quenched
		(c-f) 1200°C/1 hr – 1100°C/10 min, oil quenched, and tempered at 200°C/1 hr
25	4340	Bright field TEM micrographs
		(a-d) 1200°C/1 hr – 1100°C/30 min, oil quenched, and tempered at 200°C/1 hr.
		(e) Diffraction pattern of twinned region.
26	300M	TEM micrographs
		1200°C/1 hr and oil quenched
		(a-c) Bright field
		(d) Dark field
		(e) Diffraction pattern

FIGURE

FIGURE

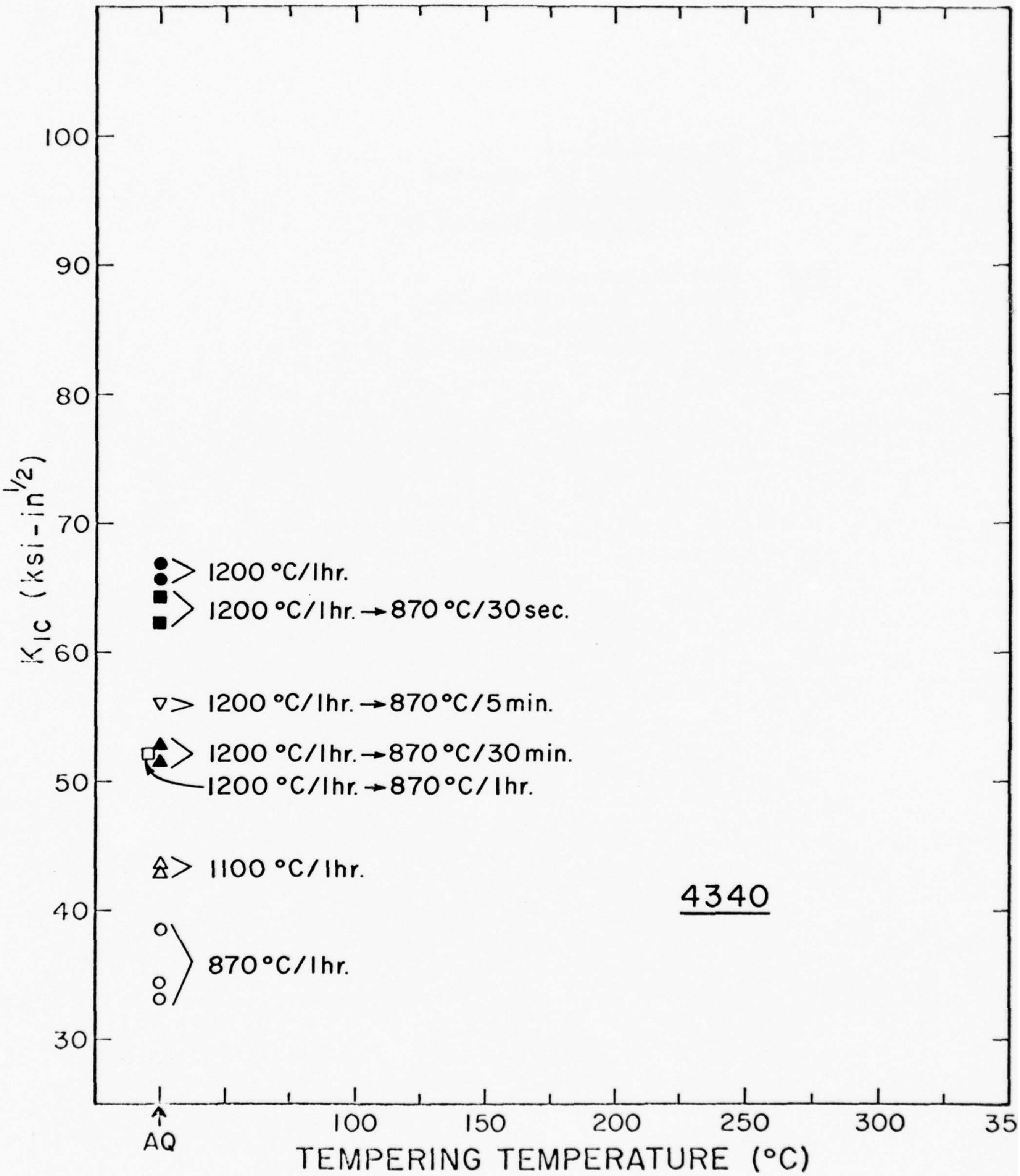


Figure 1. Fracture toughness vs. heat treatment response, tested in the as quenched condition.

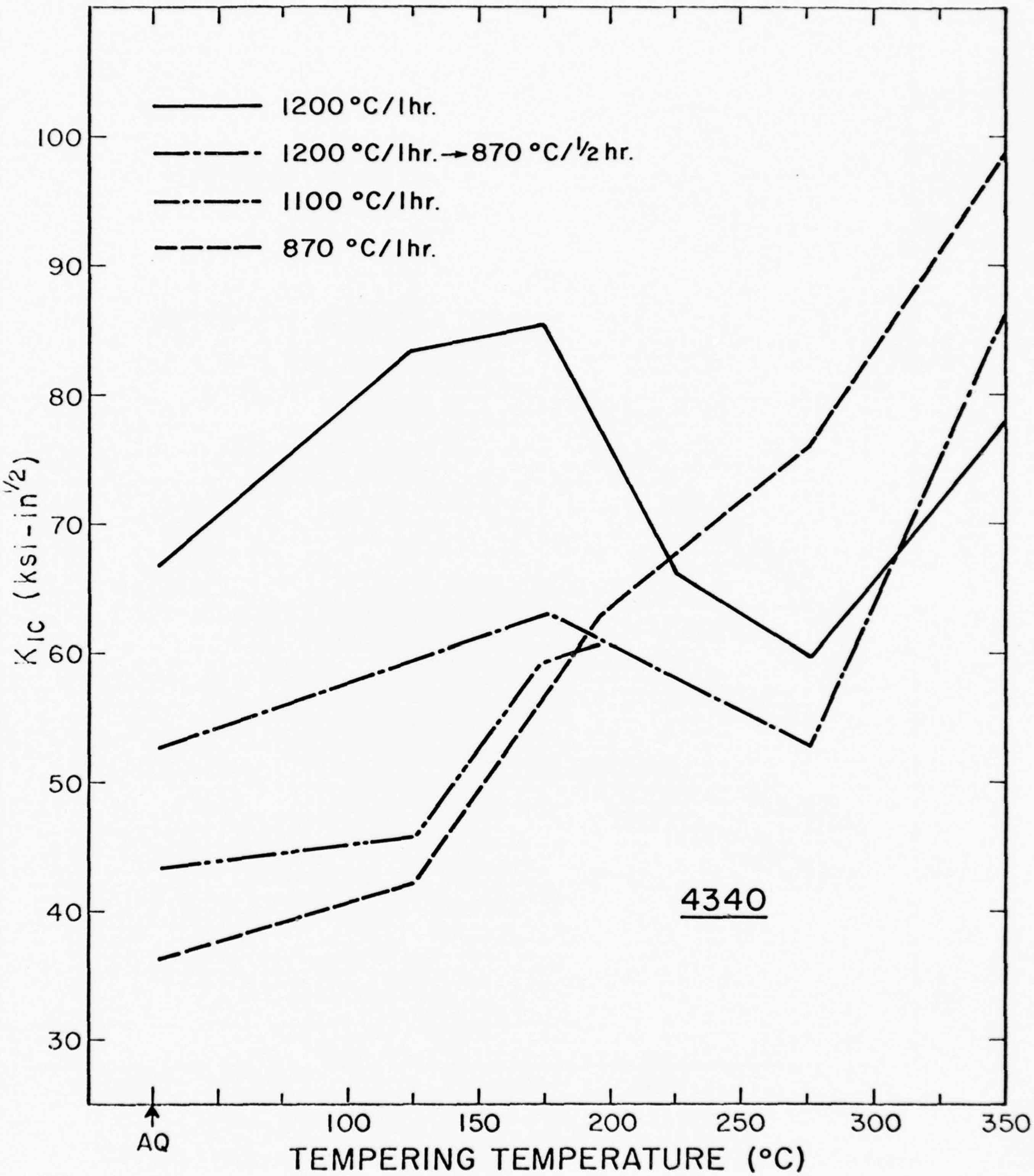


Figure 2. Fracture toughness versus tempering temperature for alloy 4340.

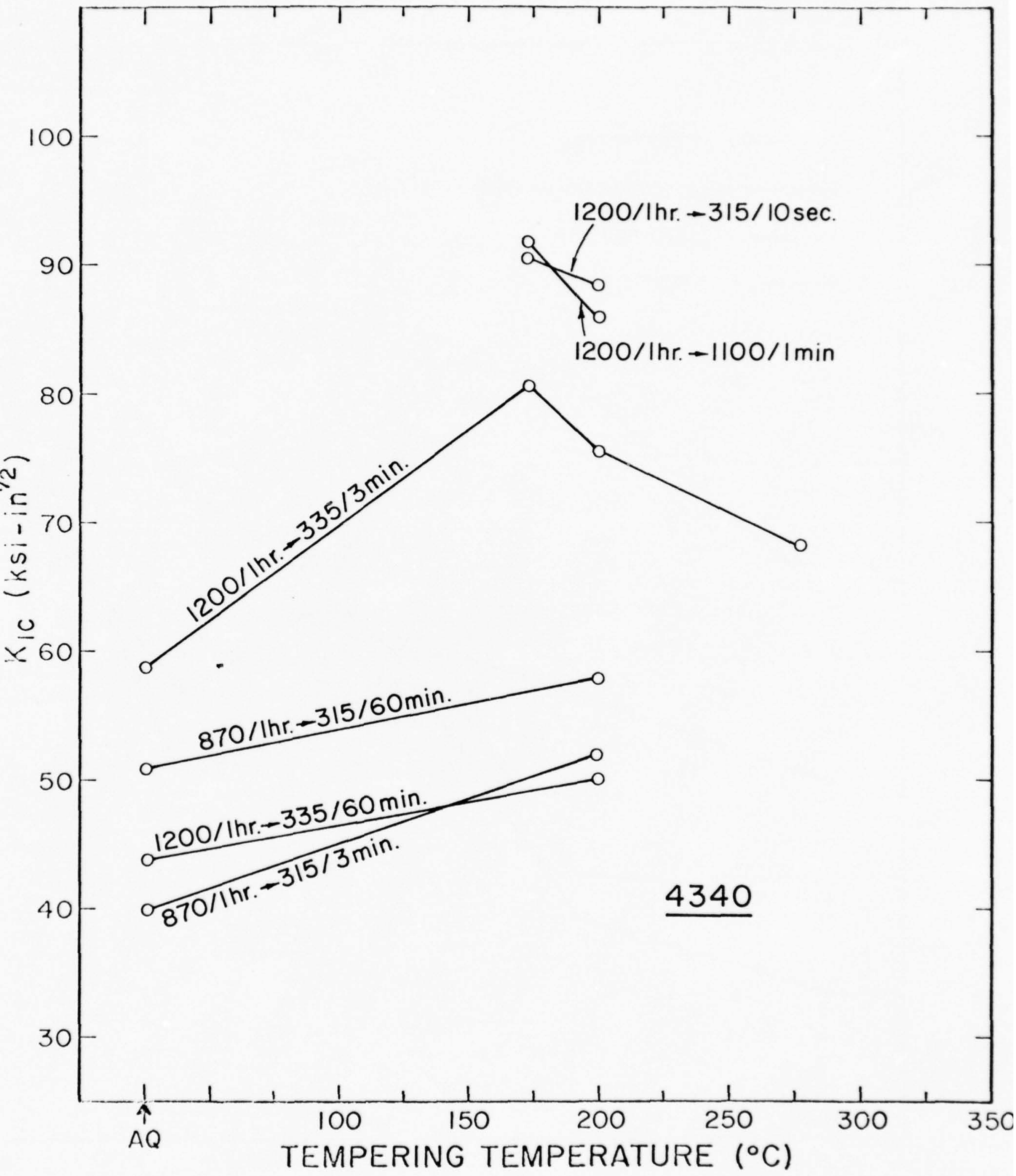


Figure 3. Fracture toughness versus tempering temperature for alloy 4340.

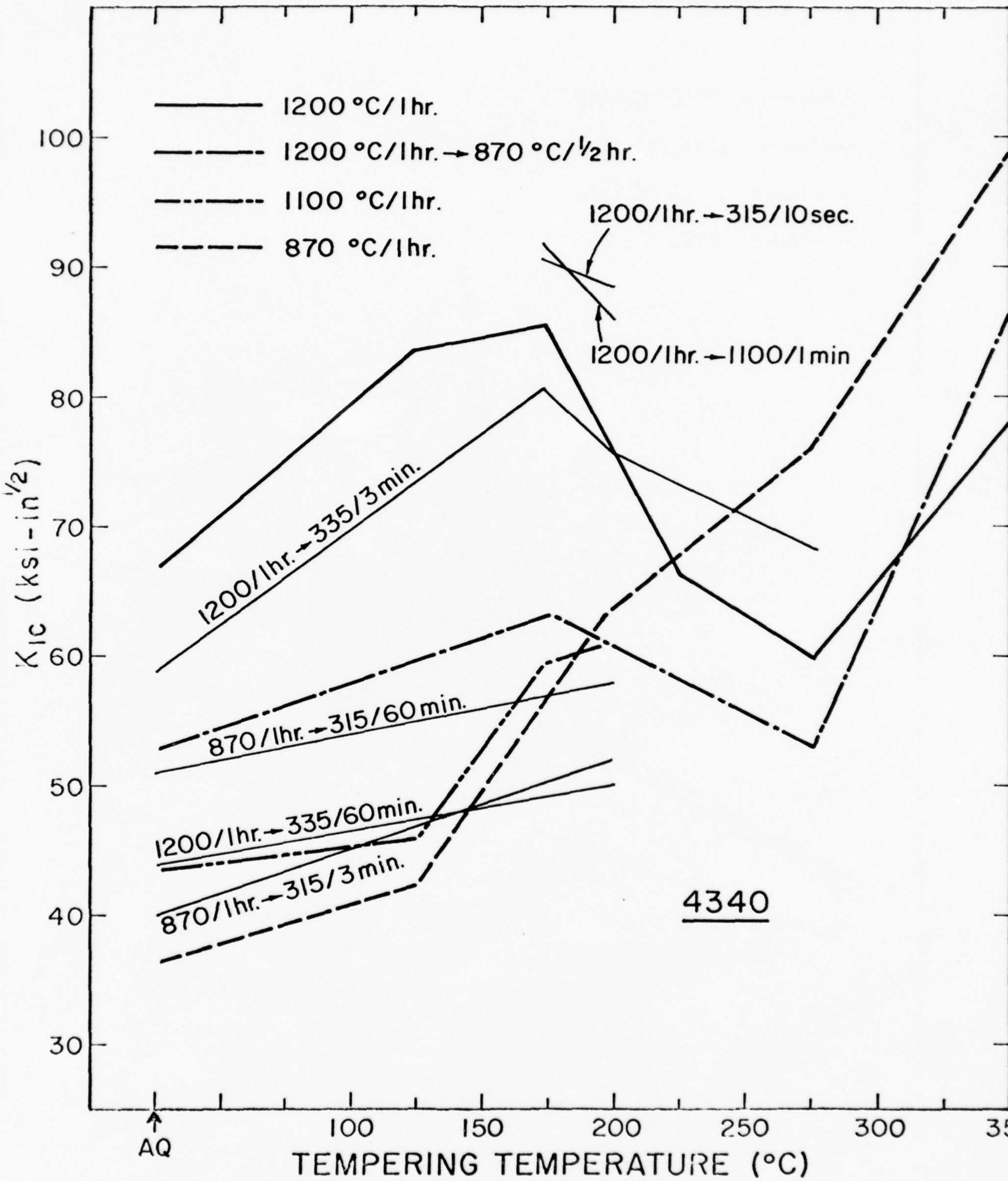


Figure 4. Composite of Figures 2 and 3.

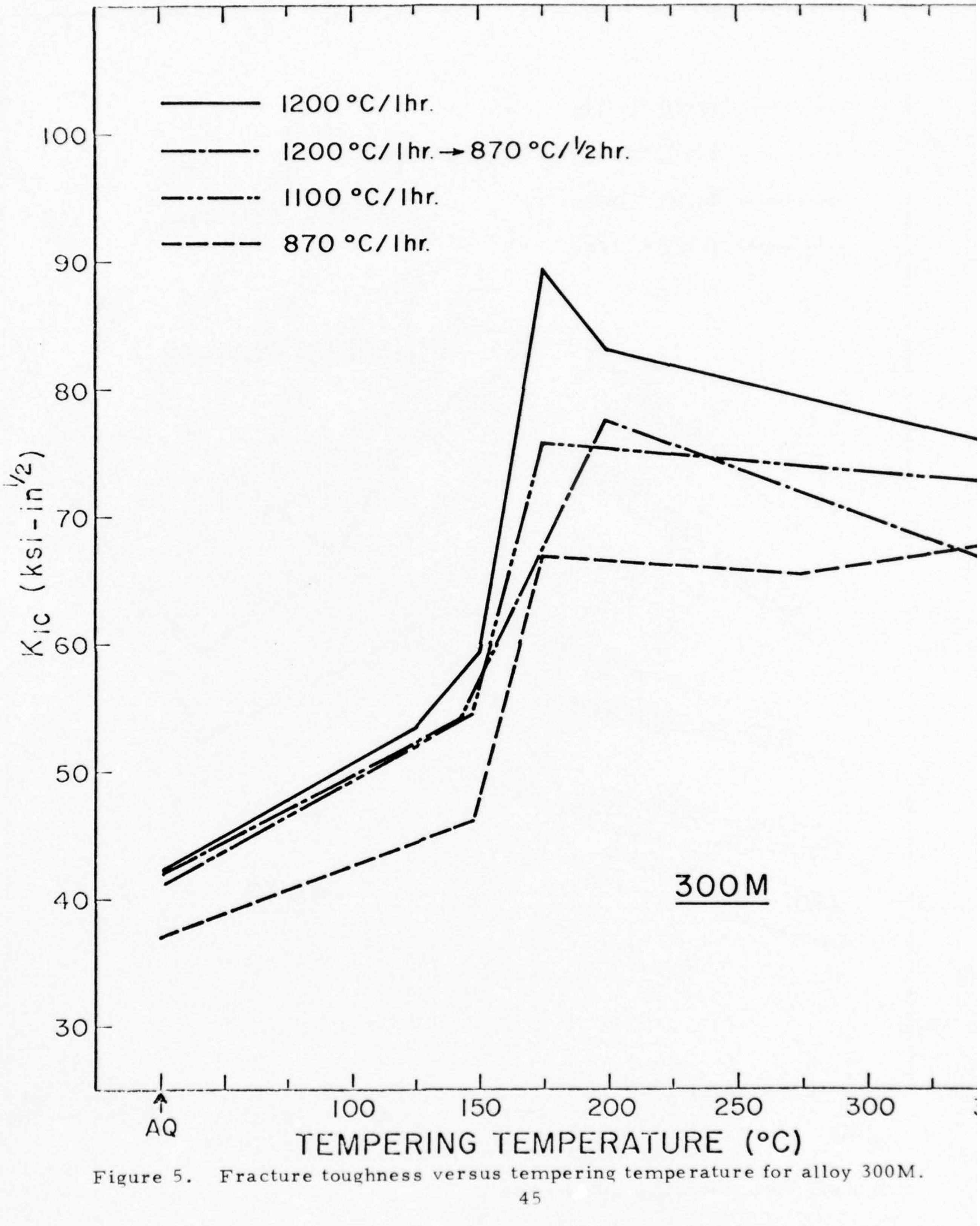


Figure 5. Fracture toughness versus tempering temperature for alloy 300M.

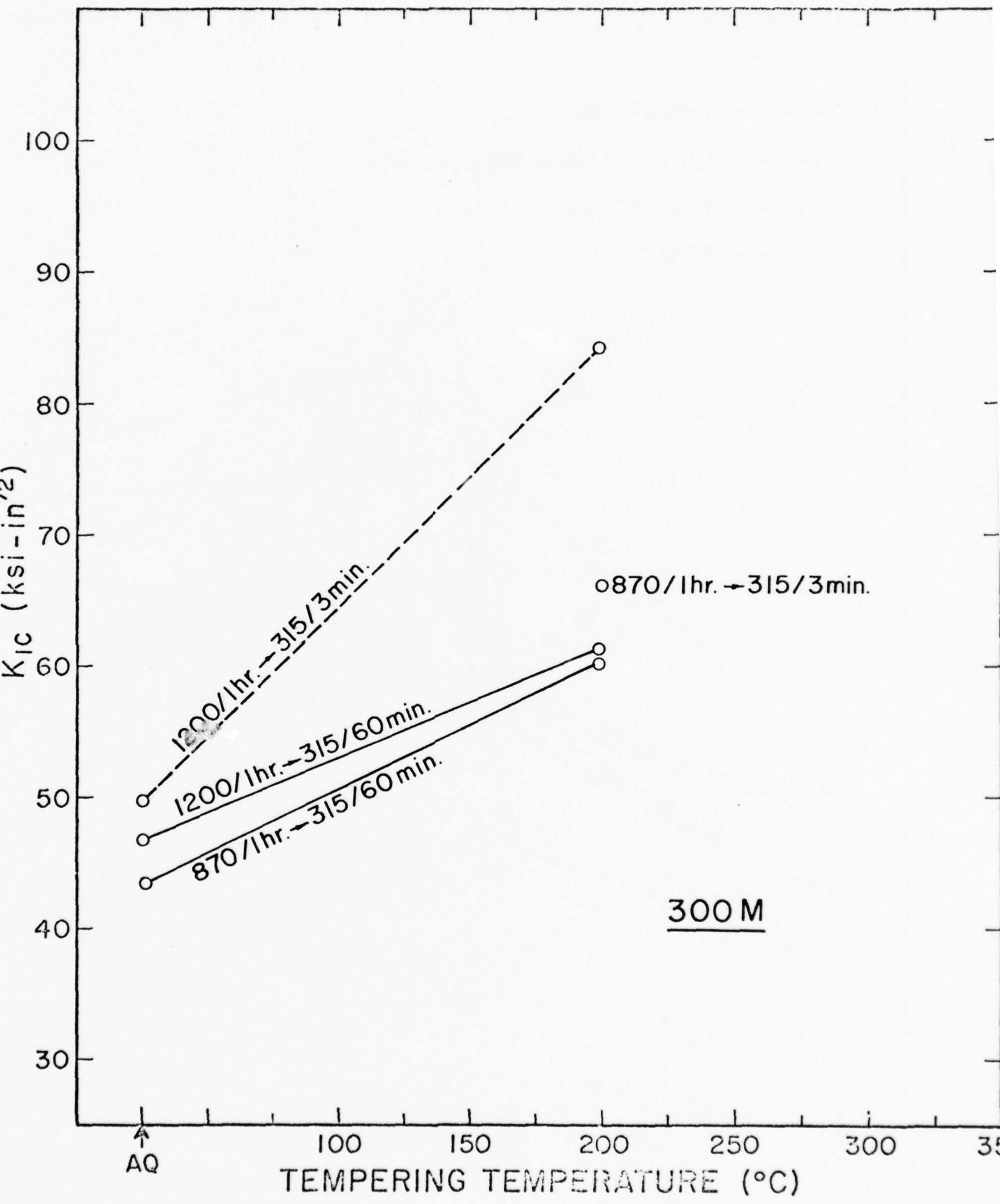


Figure 6. Fracture toughness versus tempering temperature for alloy 300M.

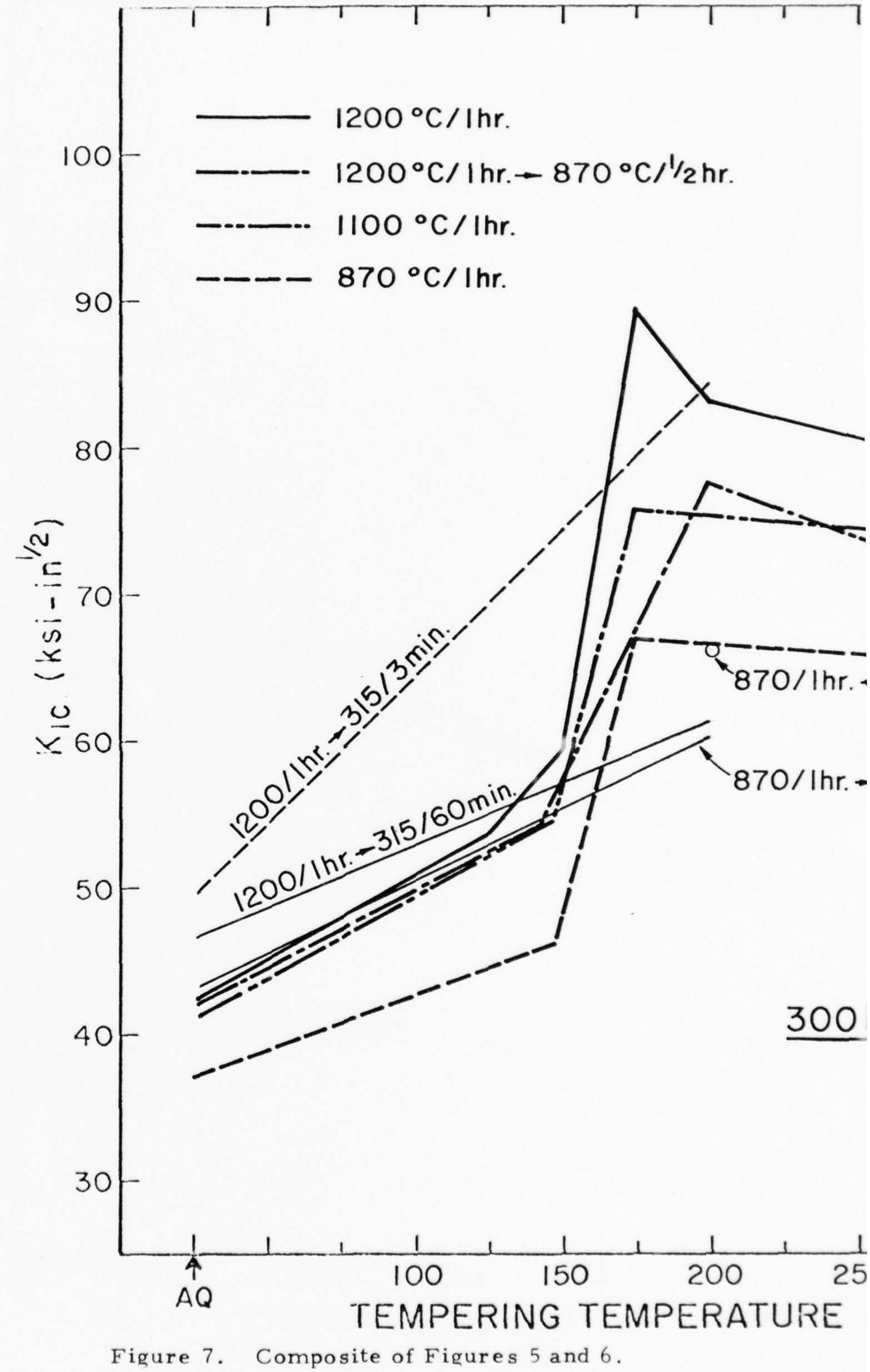


Figure 7. Composite of Figures 5 and 6.

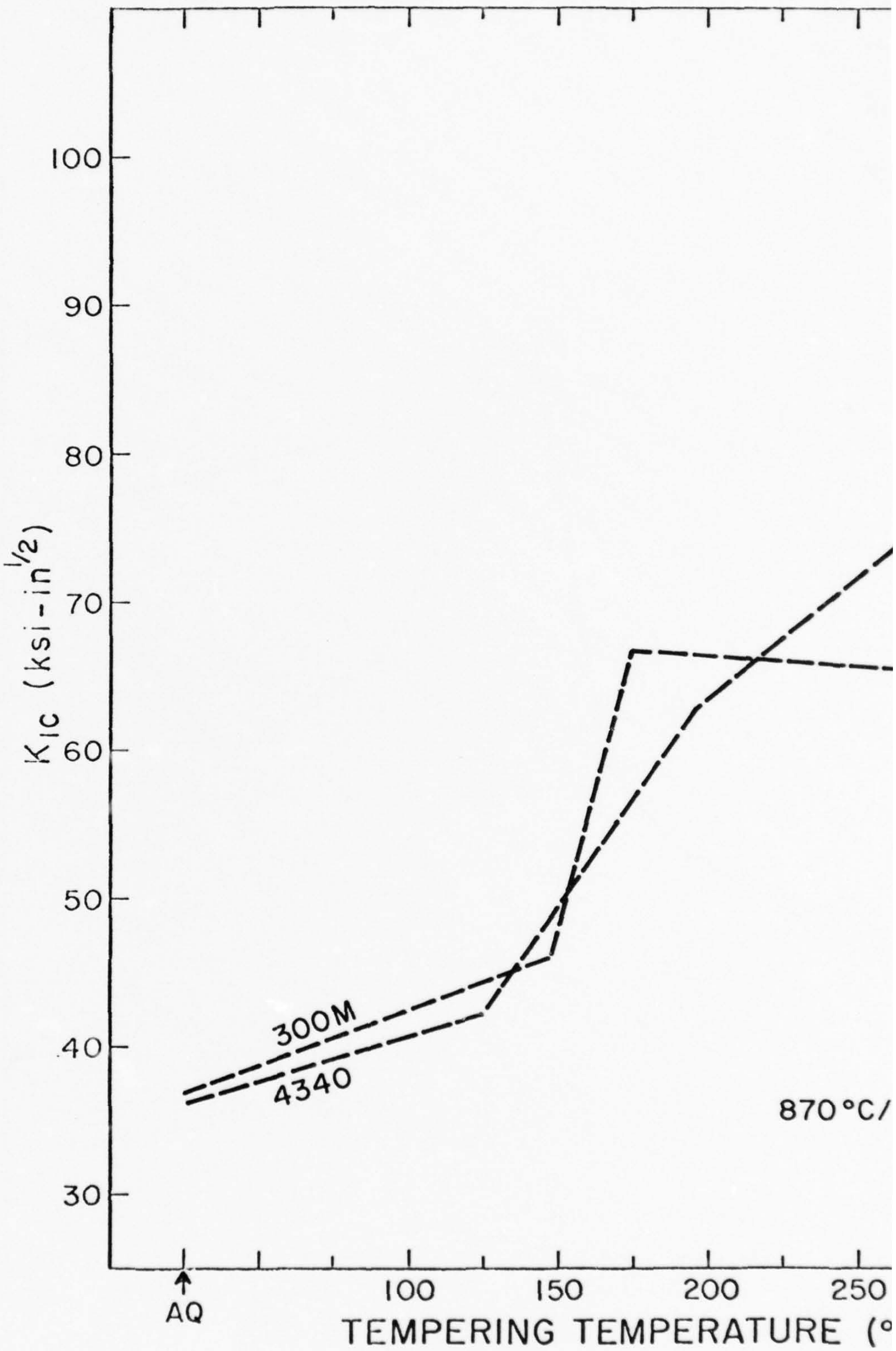


Figure 8. Fracture toughness of 300M and 4340 after a standar
solutioning treatment.

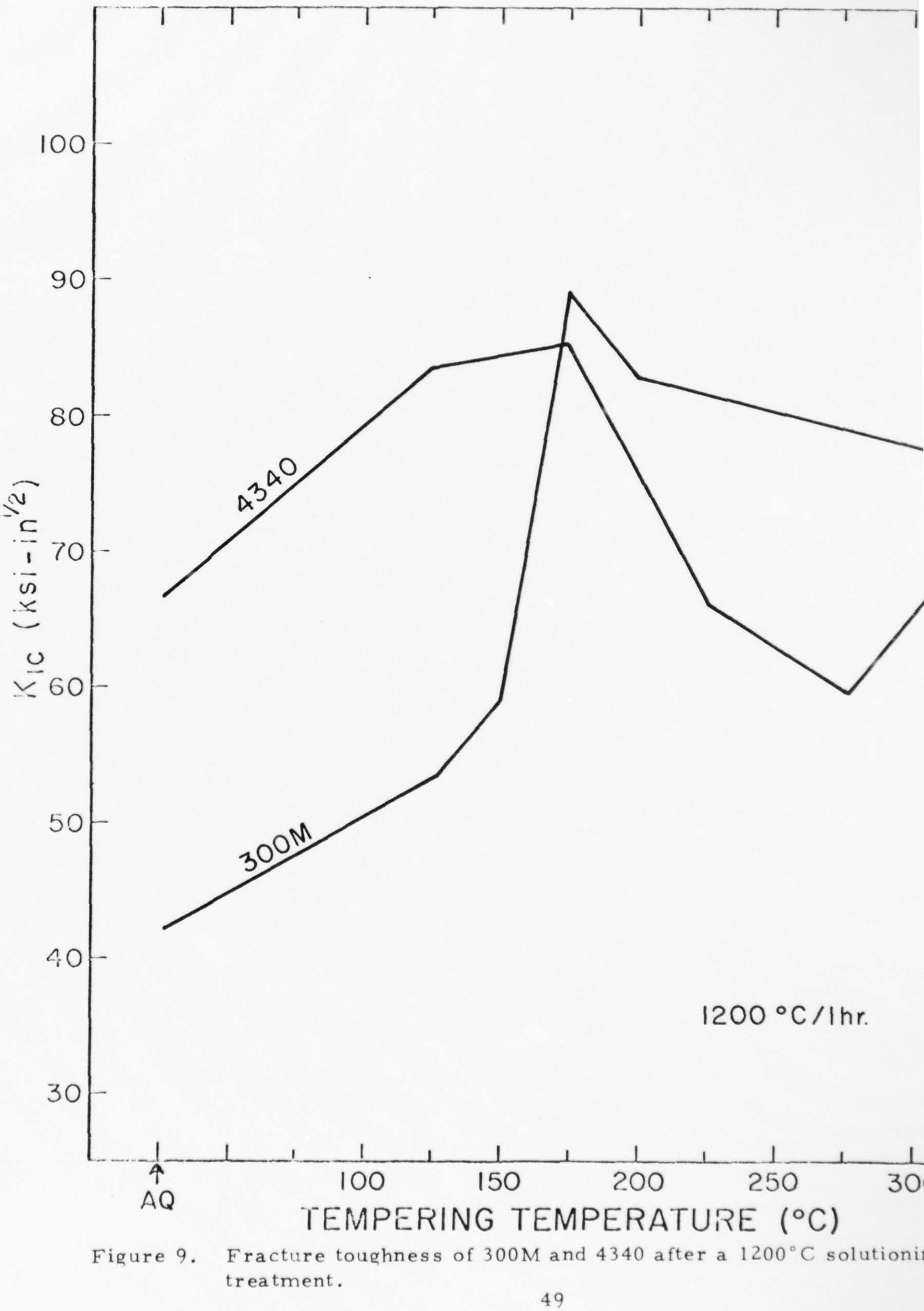


Figure 9. Fracture toughness of 300M and 4340 after a 1200°C solutionin treatment.

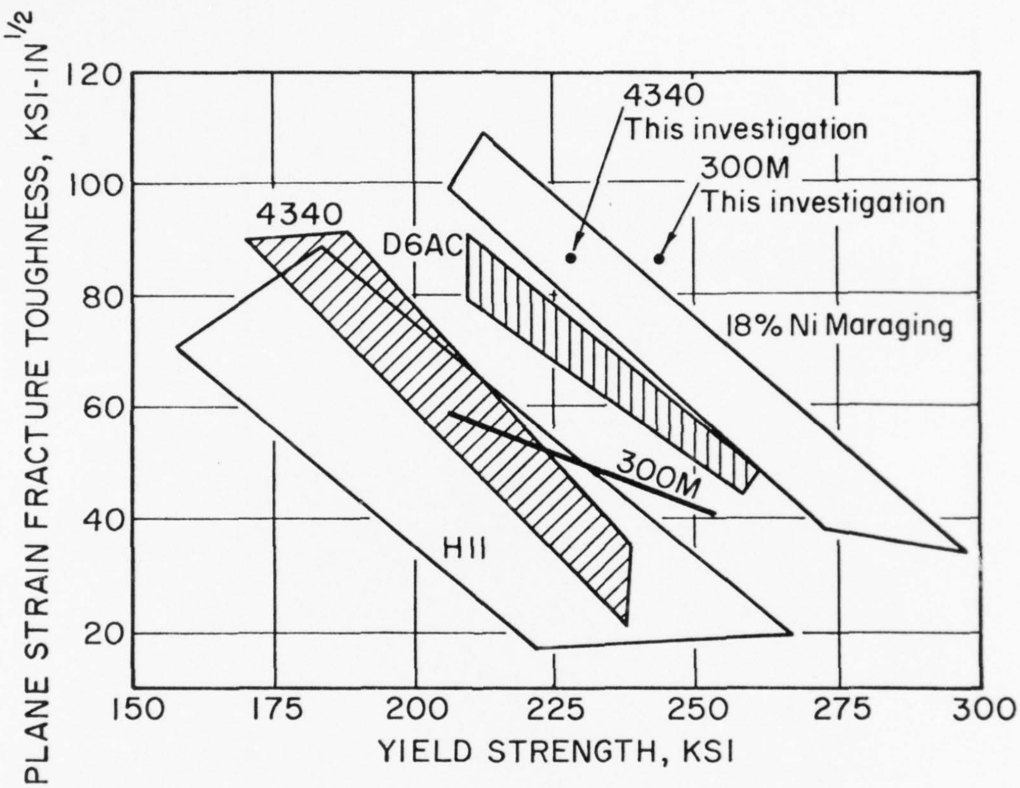


Figure 10. Yield Strength vs. plane strain fracture toughness for several commercial alloy steels.

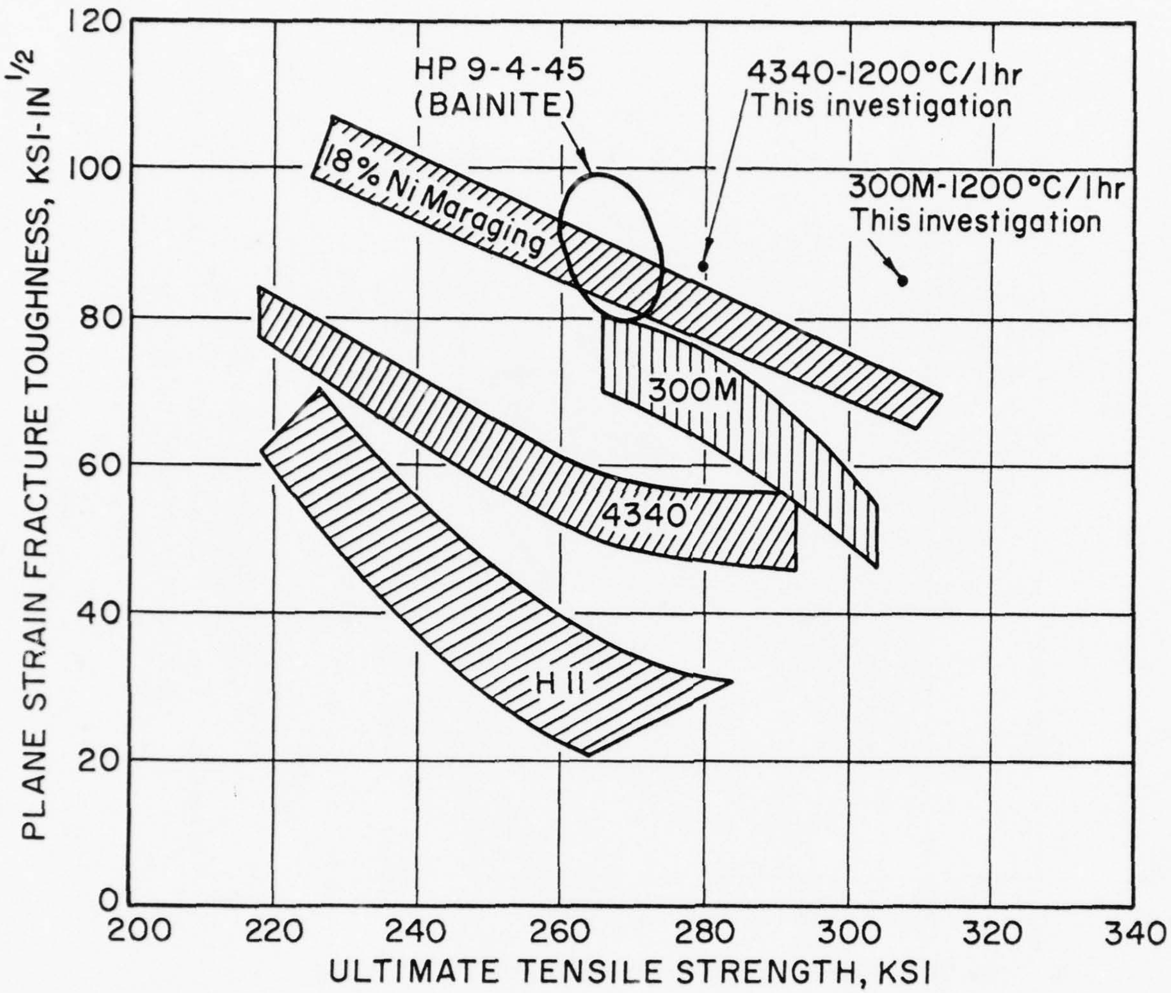


FIGURE 11. Ultimate tensile strength vs plane strain fracture toughness for several commercial alloy steels.

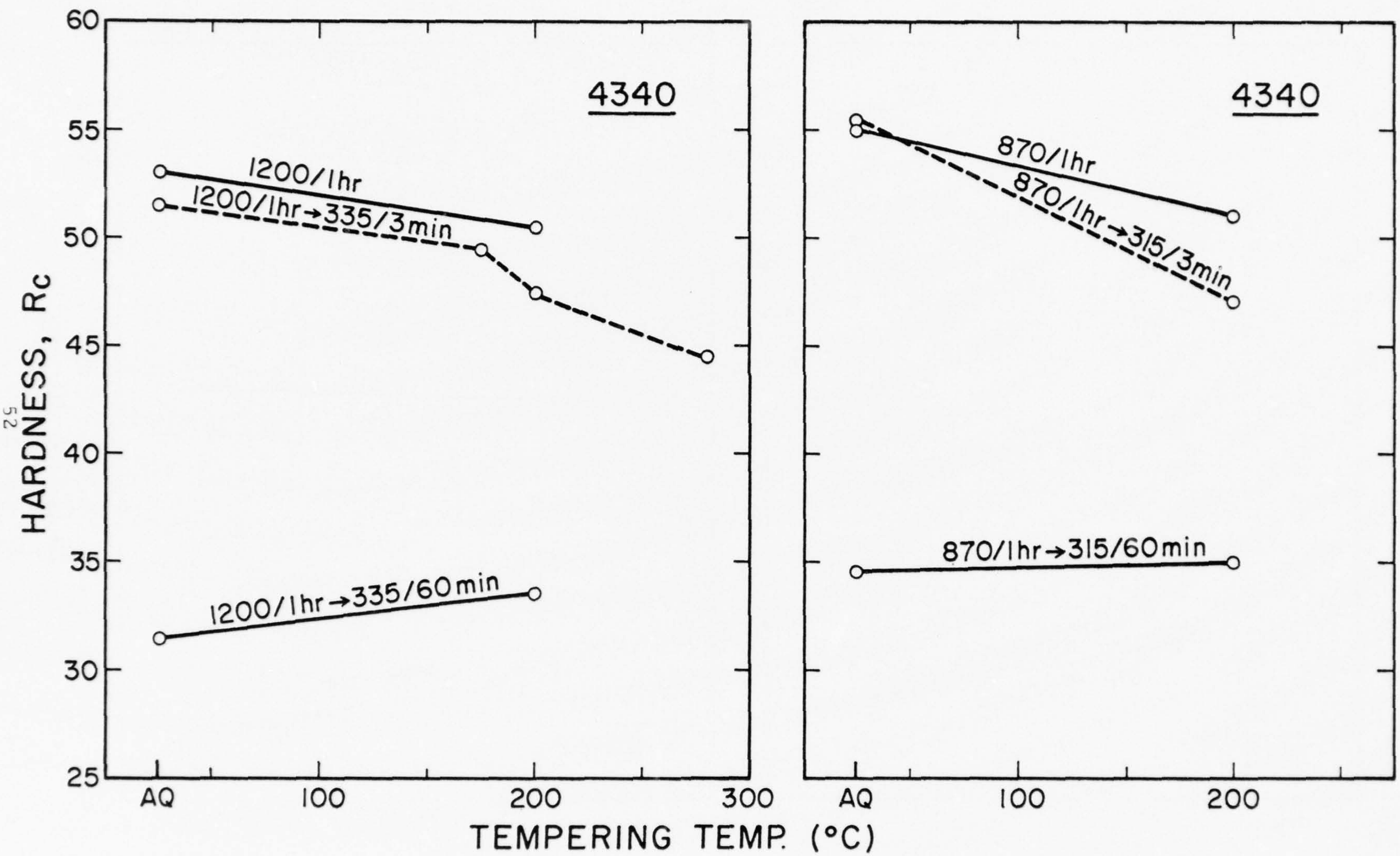


Figure 12. Hardness vs tempering temperature for alloy 4340.

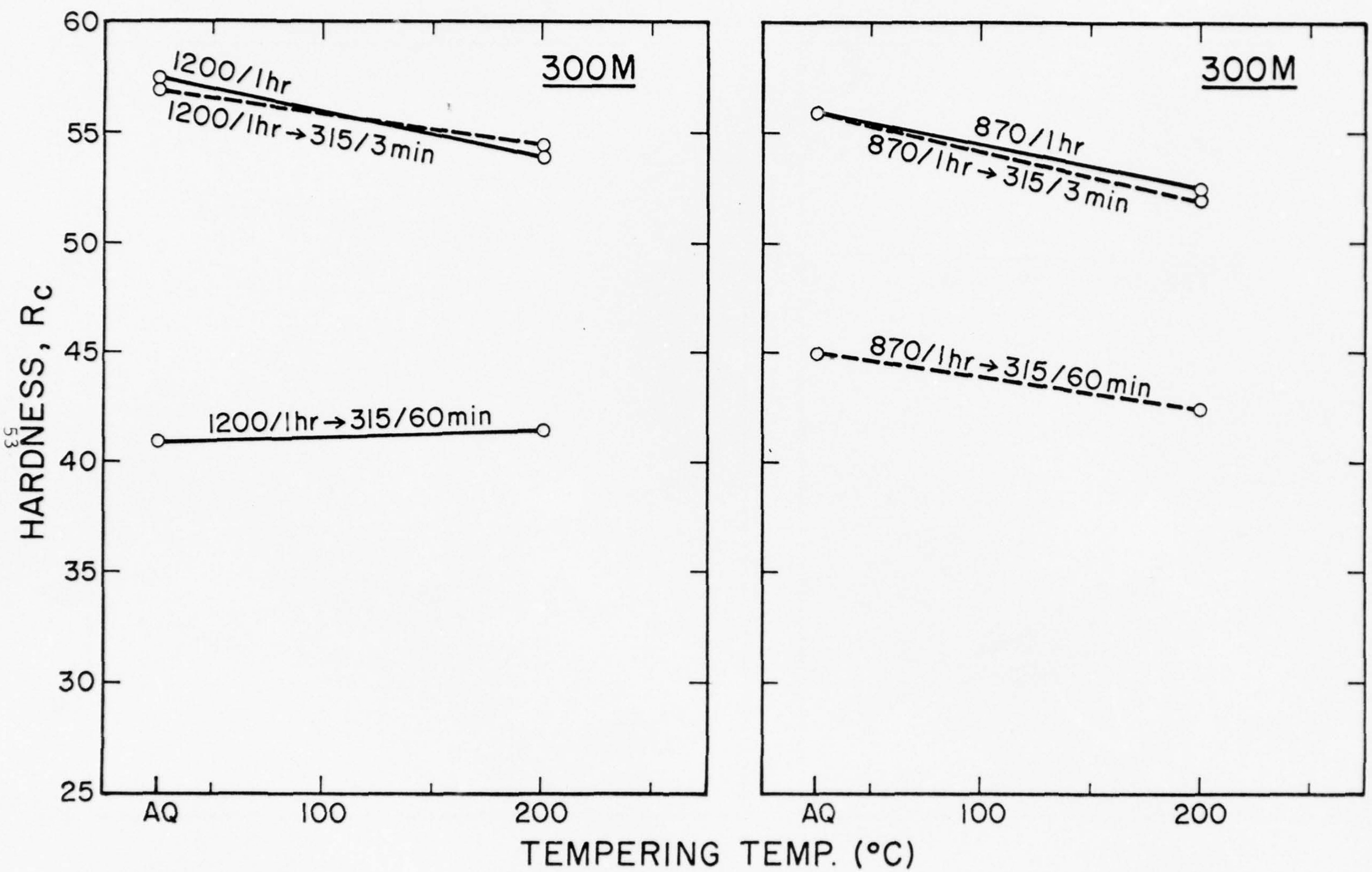


Figure 13. Hardness vs tempering temperature for alloy 300M.

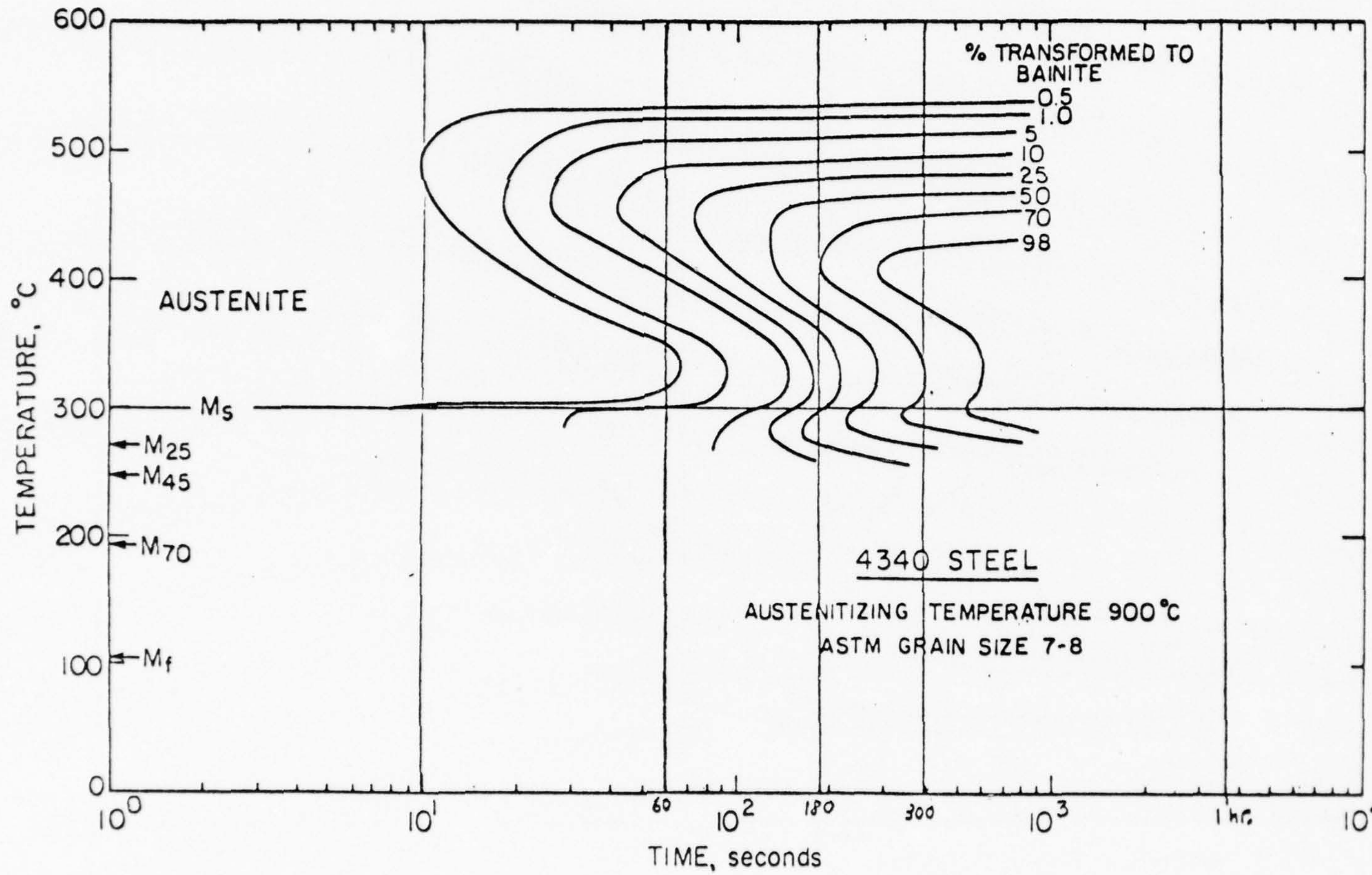


Figure 14. Isothermal transformation curve for alloy 4340 austenitized at 900°C

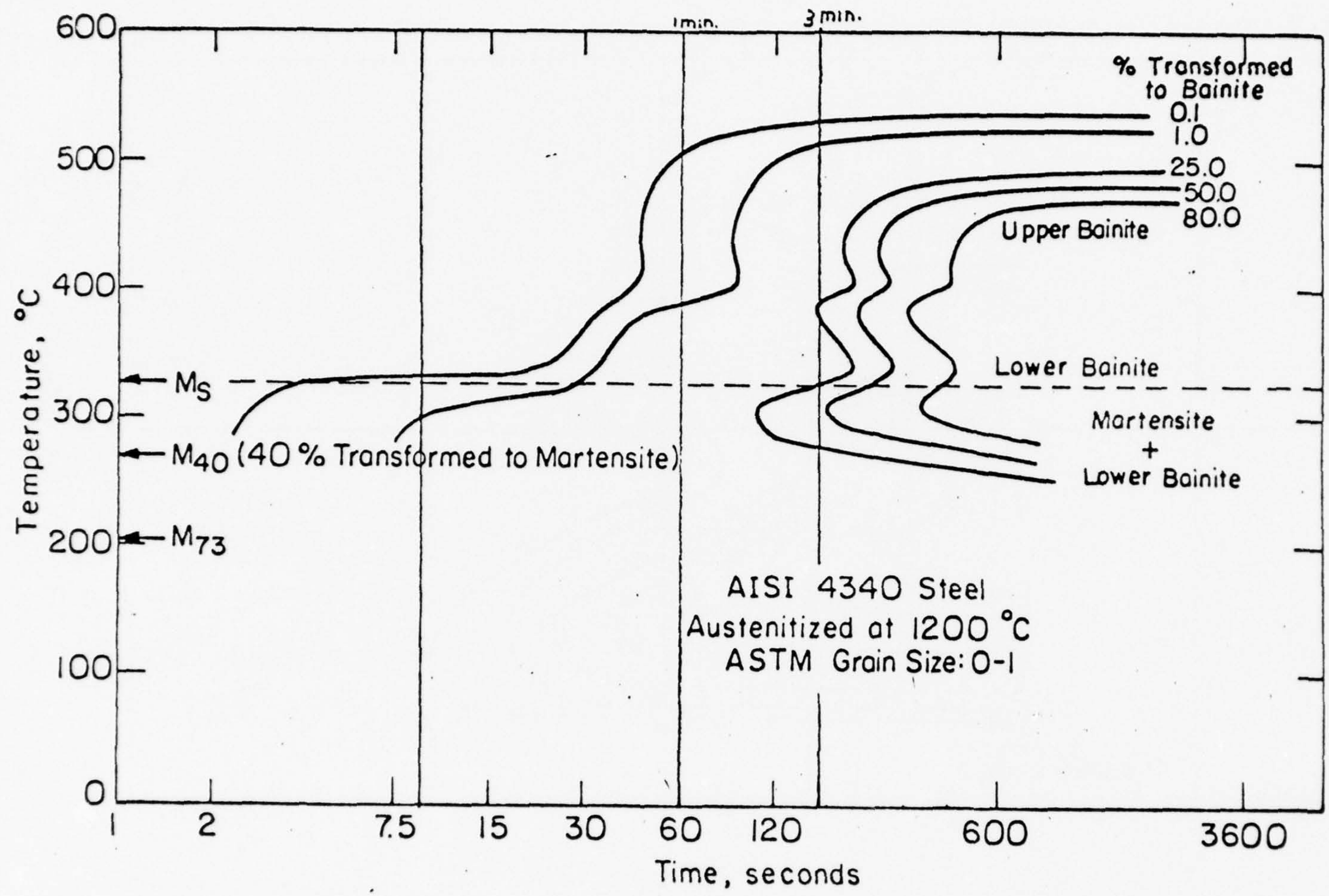


Figure 15. Isothermal transformation curve for alloy 4340 austenitized at 1200°C

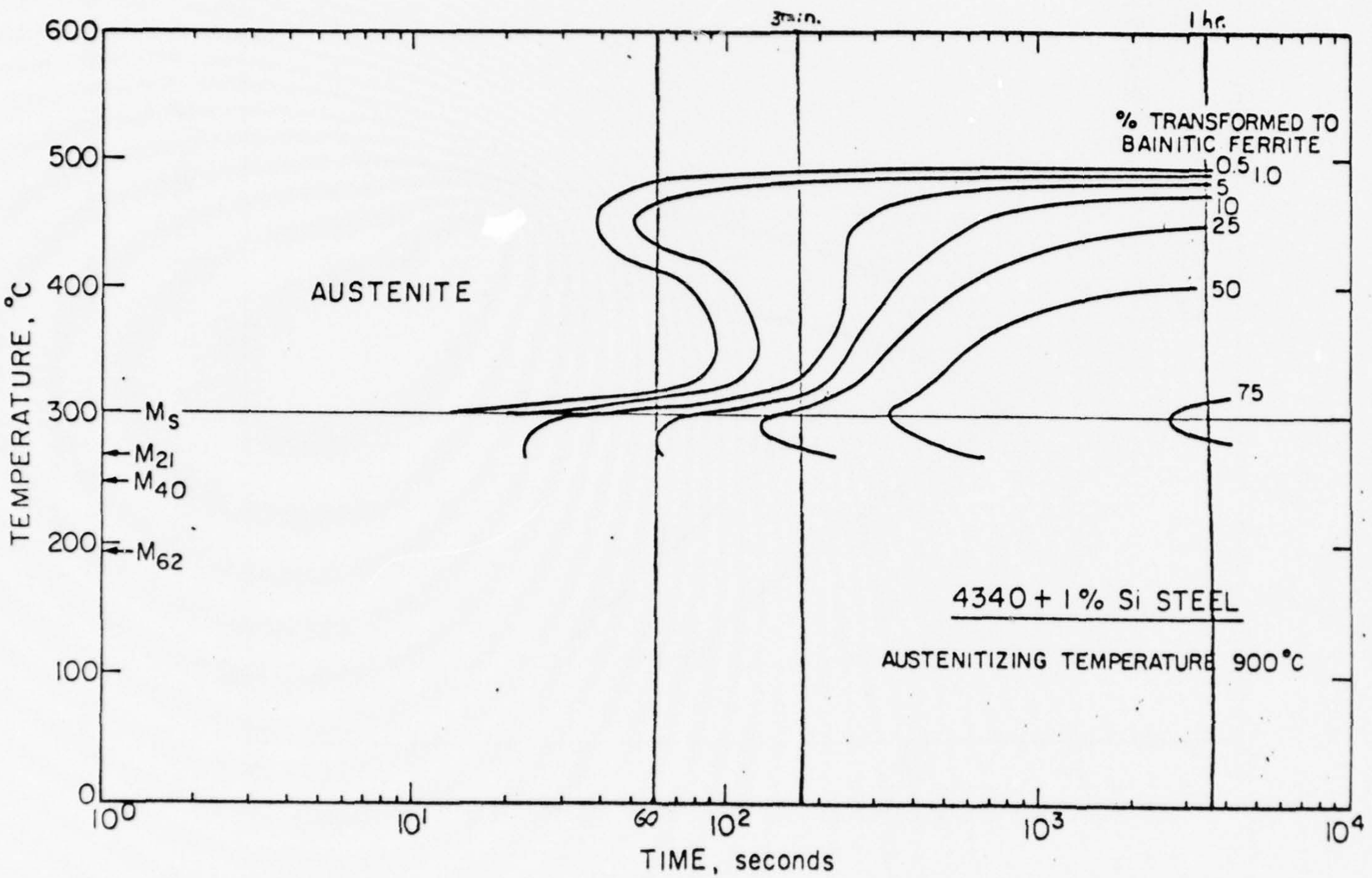


Figure 16. Isothermal transformation curve for alloy 4340 + 1% Si austenitized at 900°C

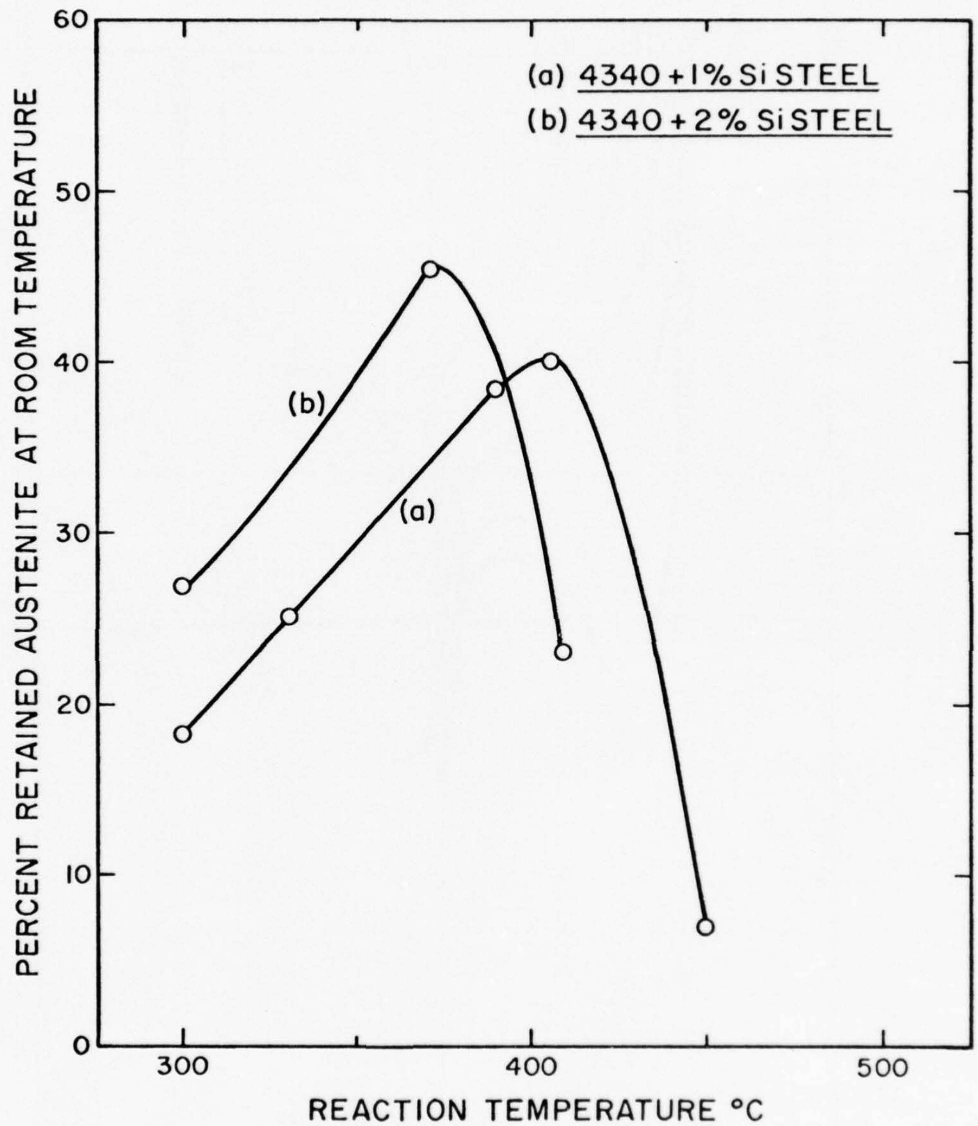
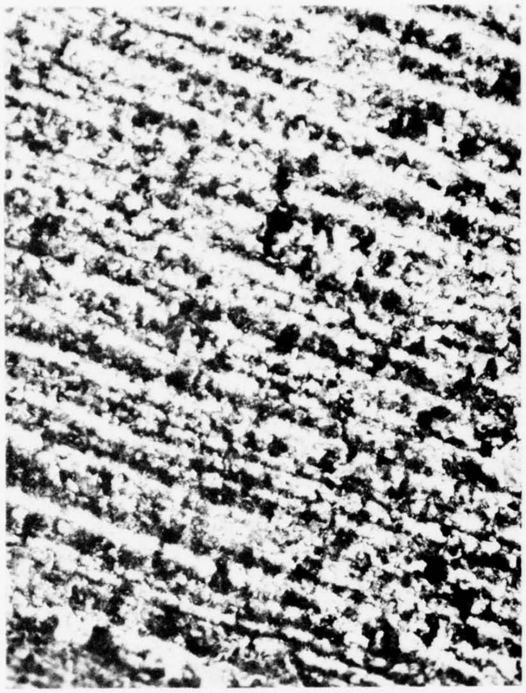
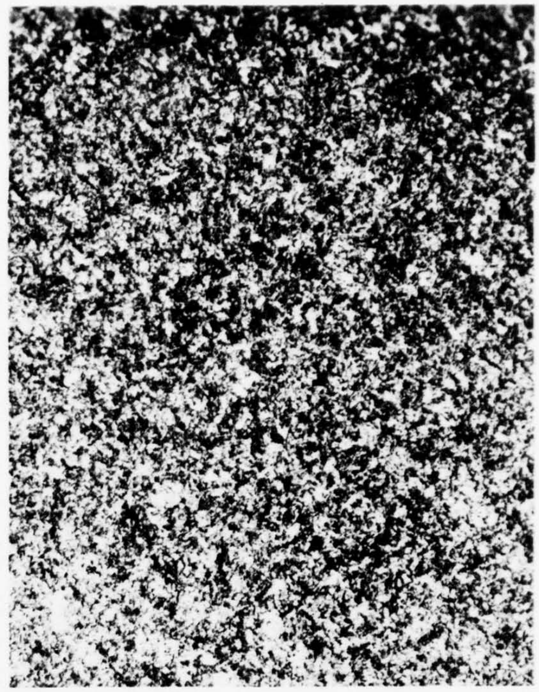


Figure 17. Reaction temperature vs. percent retained austenite for Si modified 4340 steel.

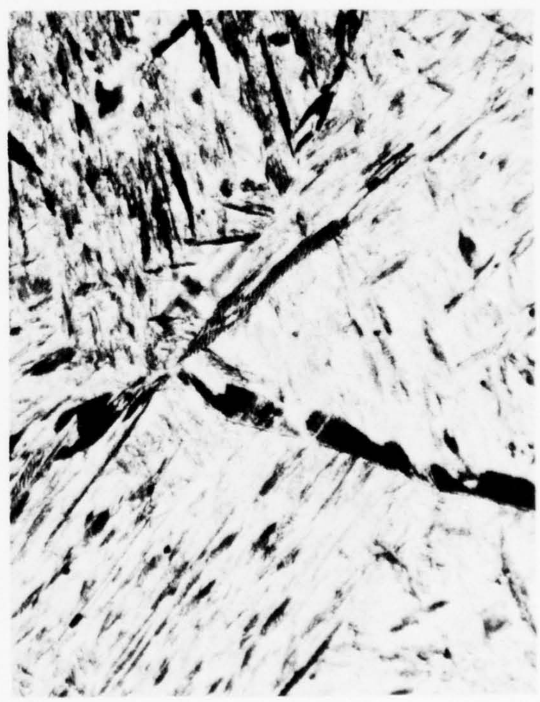


(a) 100X



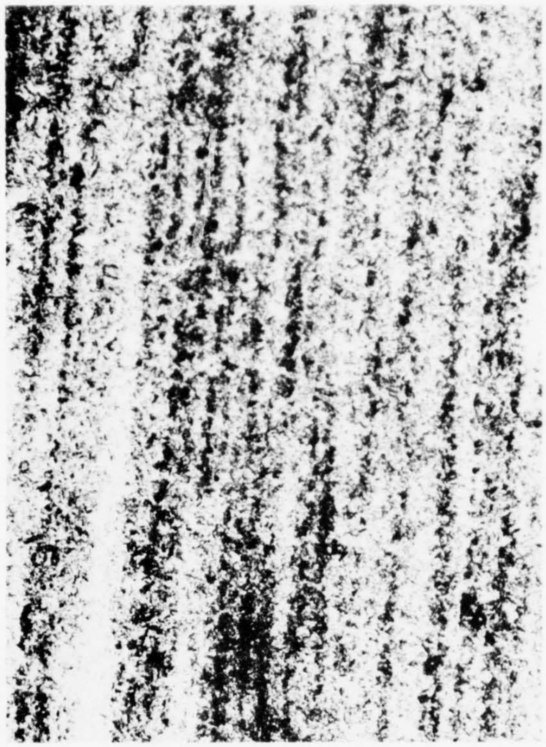
(b) 100X

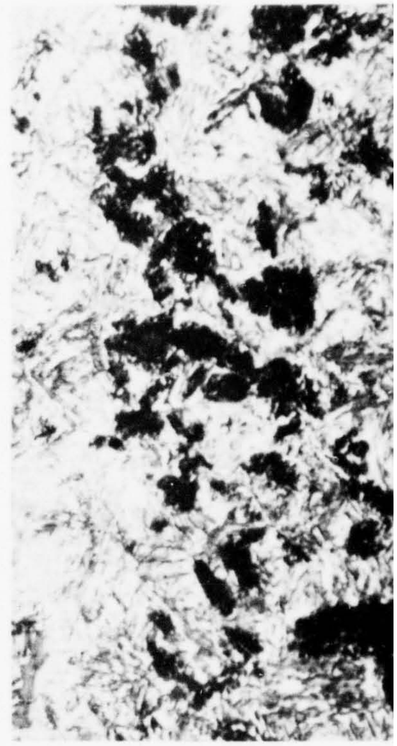
(c) 100X



(d) 1000X

Figure 18.





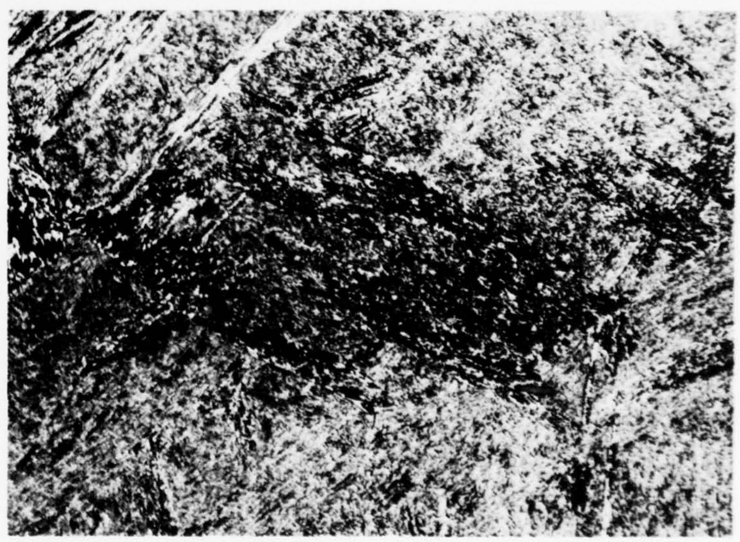
(a) 100X (b)

(c) 4000X

Figure 19.

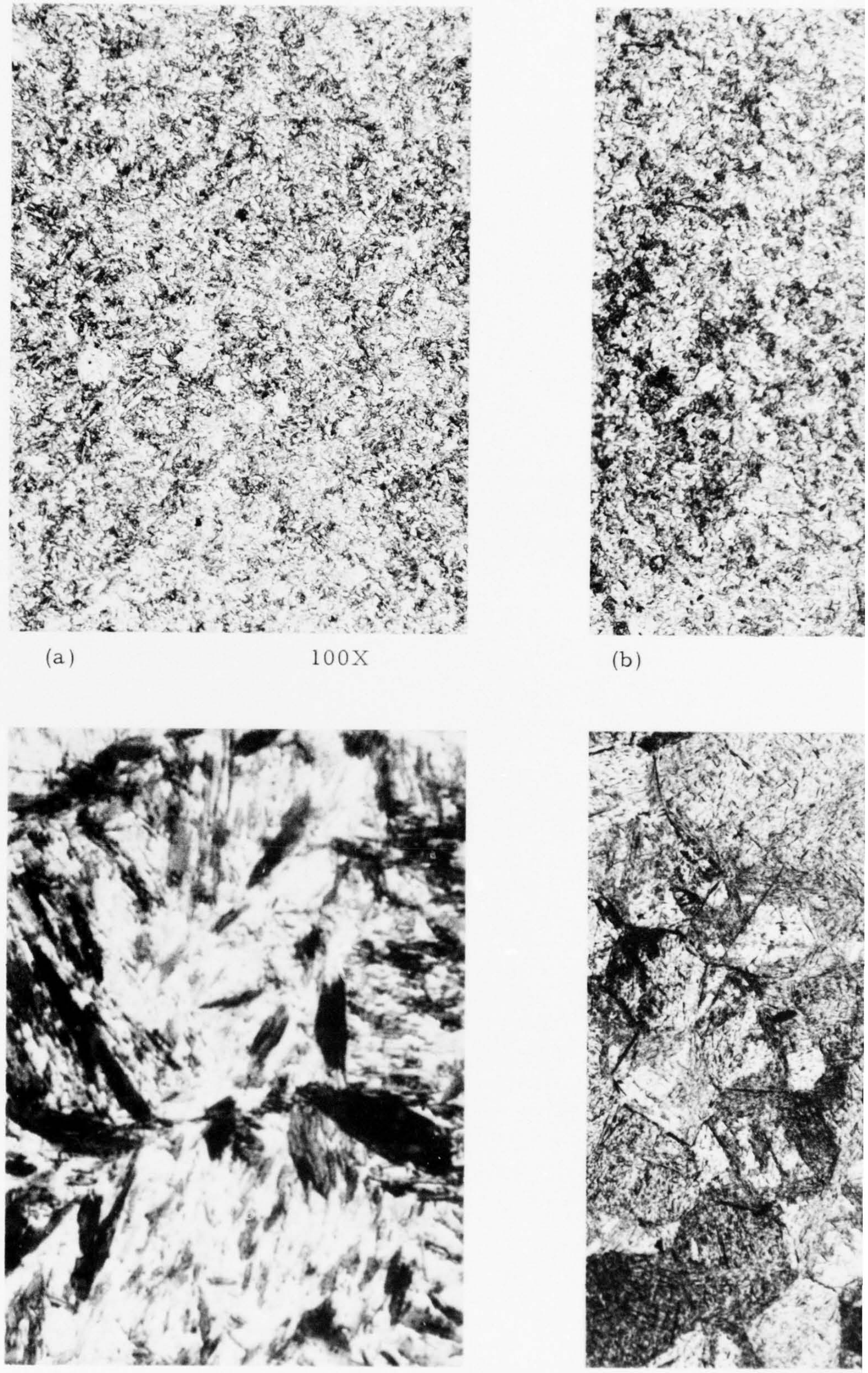
500X

(a) 500X



(b) 500X

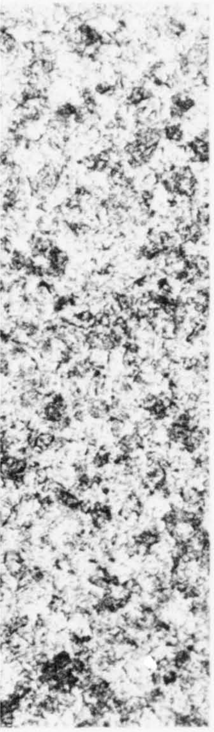
Figure 20.



(a) 100X (b)

(c) 2000X (d)

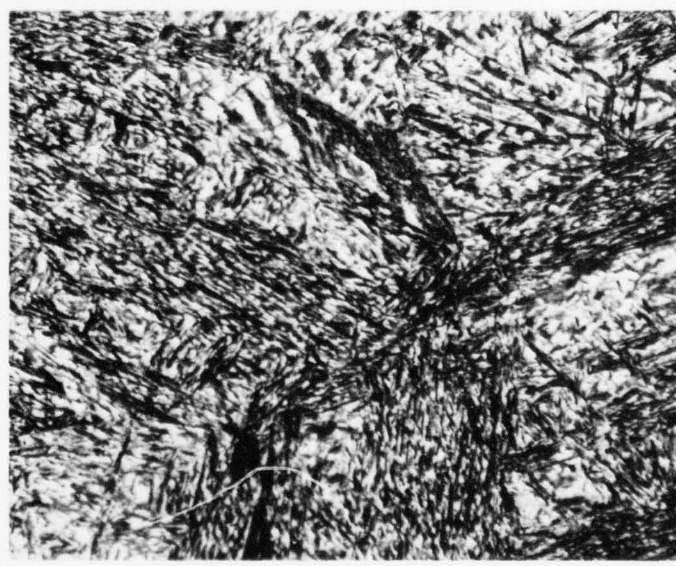
Figure 21.



100X

100X

(a) 500X

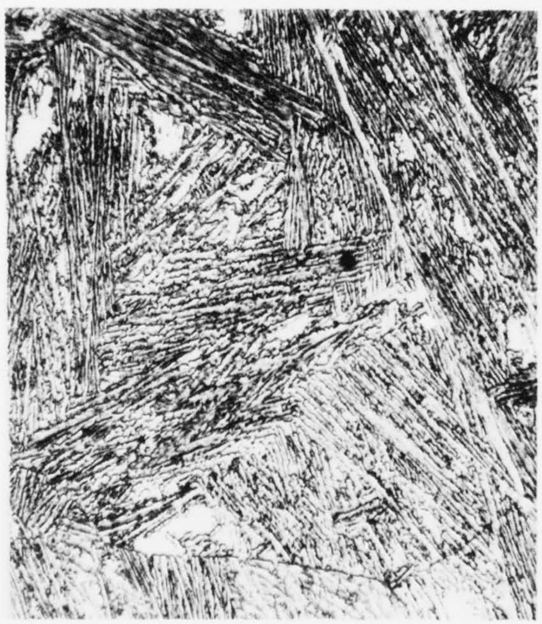


(b) 500X

Figure 22.

(a) 500X

(b) 500



(c) 500X

(d) 500

Figure 23.

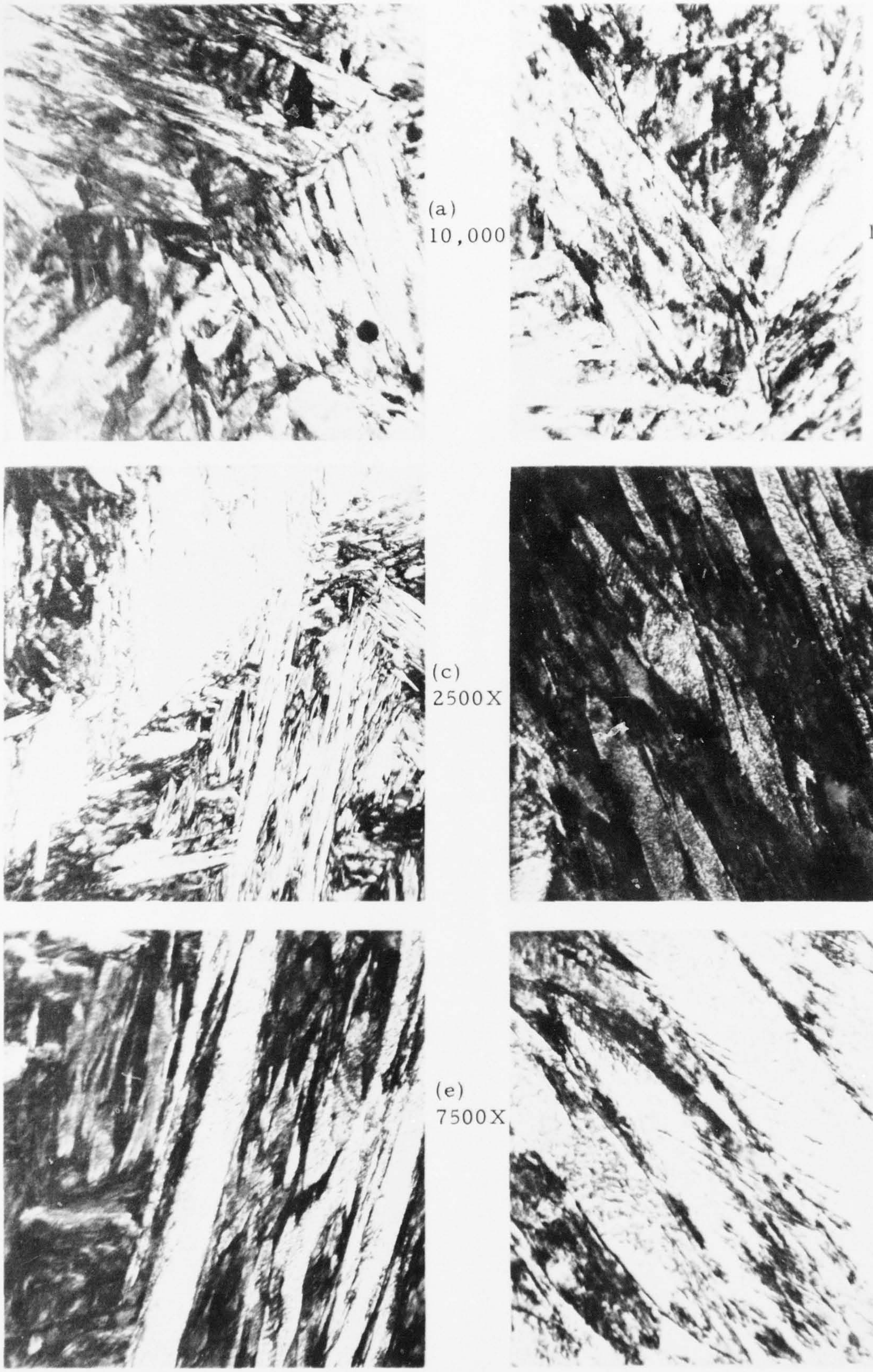
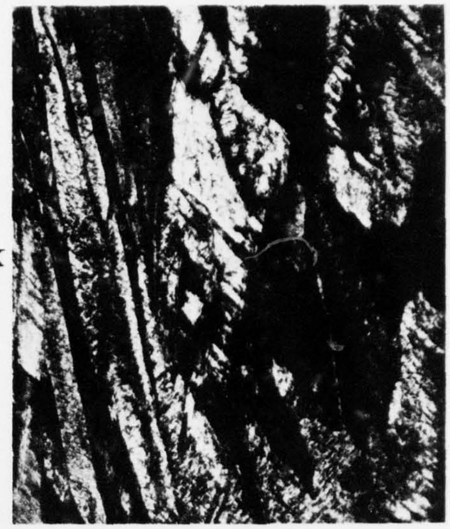


Figure 24.

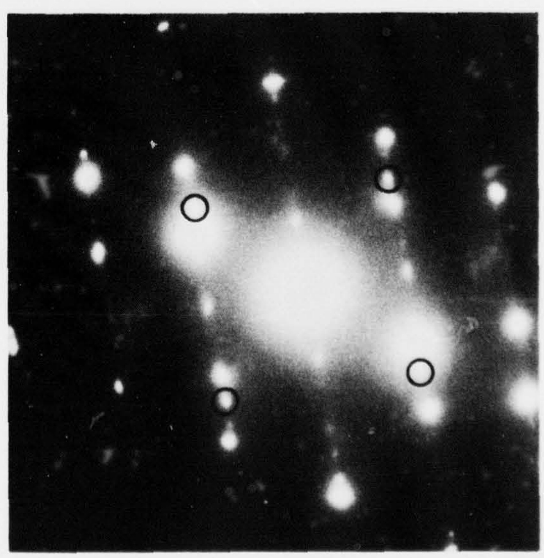
(a)
18,000X



(b)
18,000X

(c)
18,000X

(d)
30,000X



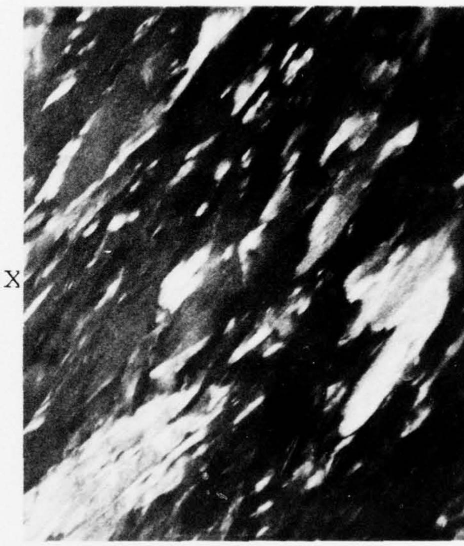
(e)

Figure 25.

(a)
10,000X

(b)
12,000X

(c)
12,000X



(d)
12,000X

(e)

Figure 26.

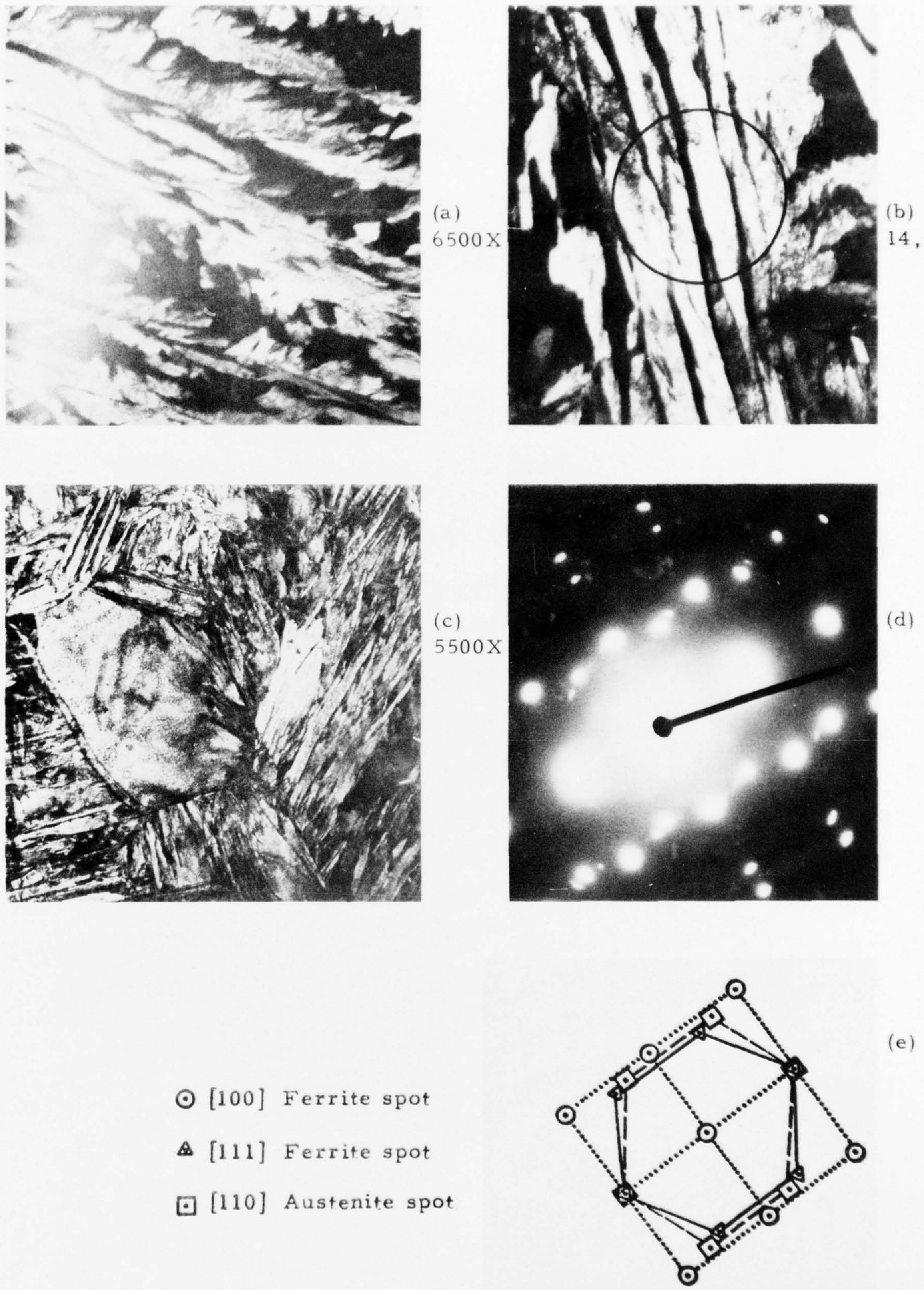
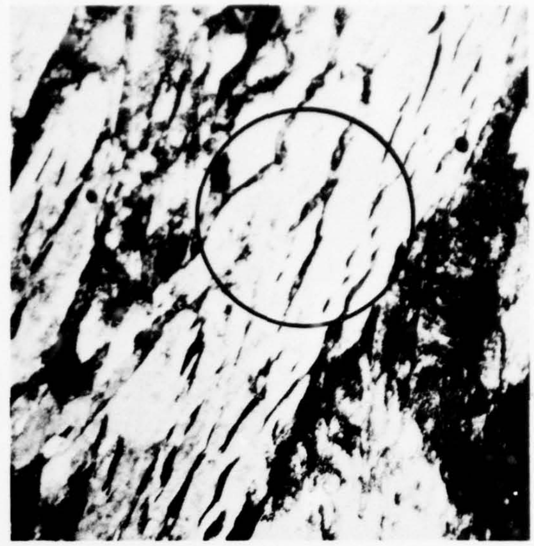
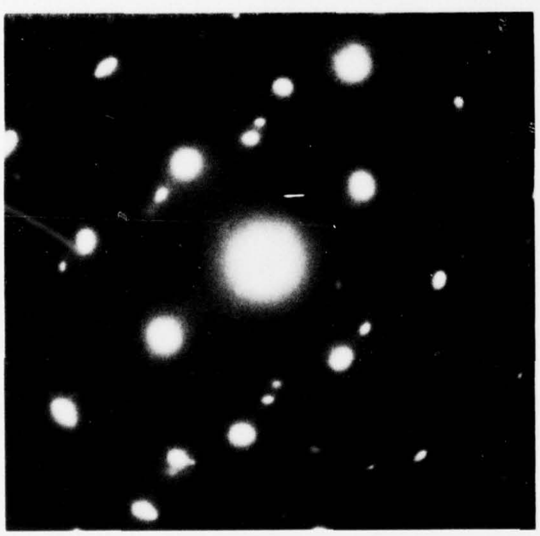

(a)
6500X
(b)
14,000X
(c)
5500X
(d)
[100] Ferrite spot
[111] Ferrite spot
[110] Austenite spot
(e)

Figure 27

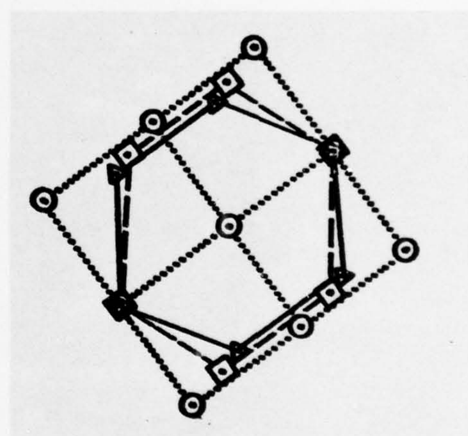
(a) 2500X



(b) 18,000X



(c)



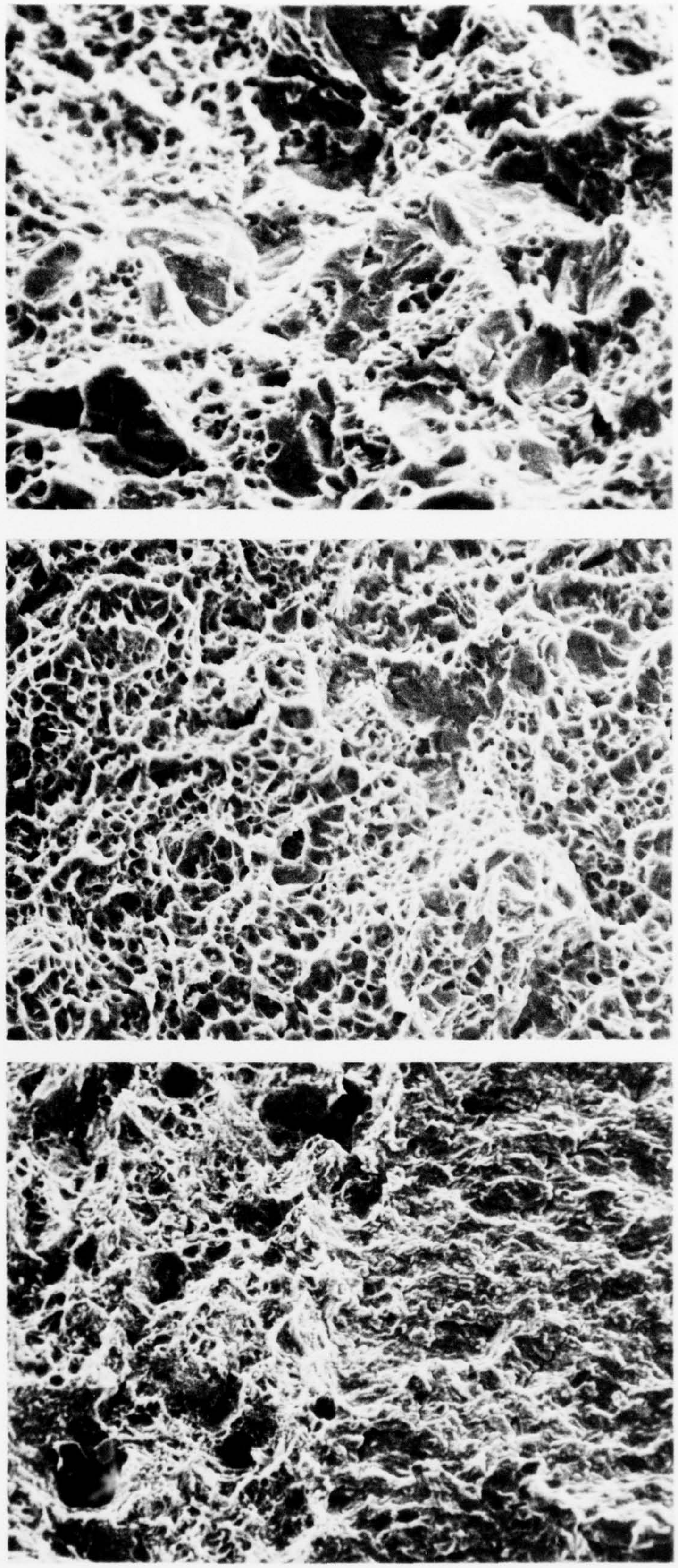
⊙ [100] Ferrite spot

▲ [111] Ferrite spot

⊡ [110] Austenite spot

(d)

Figure 28.

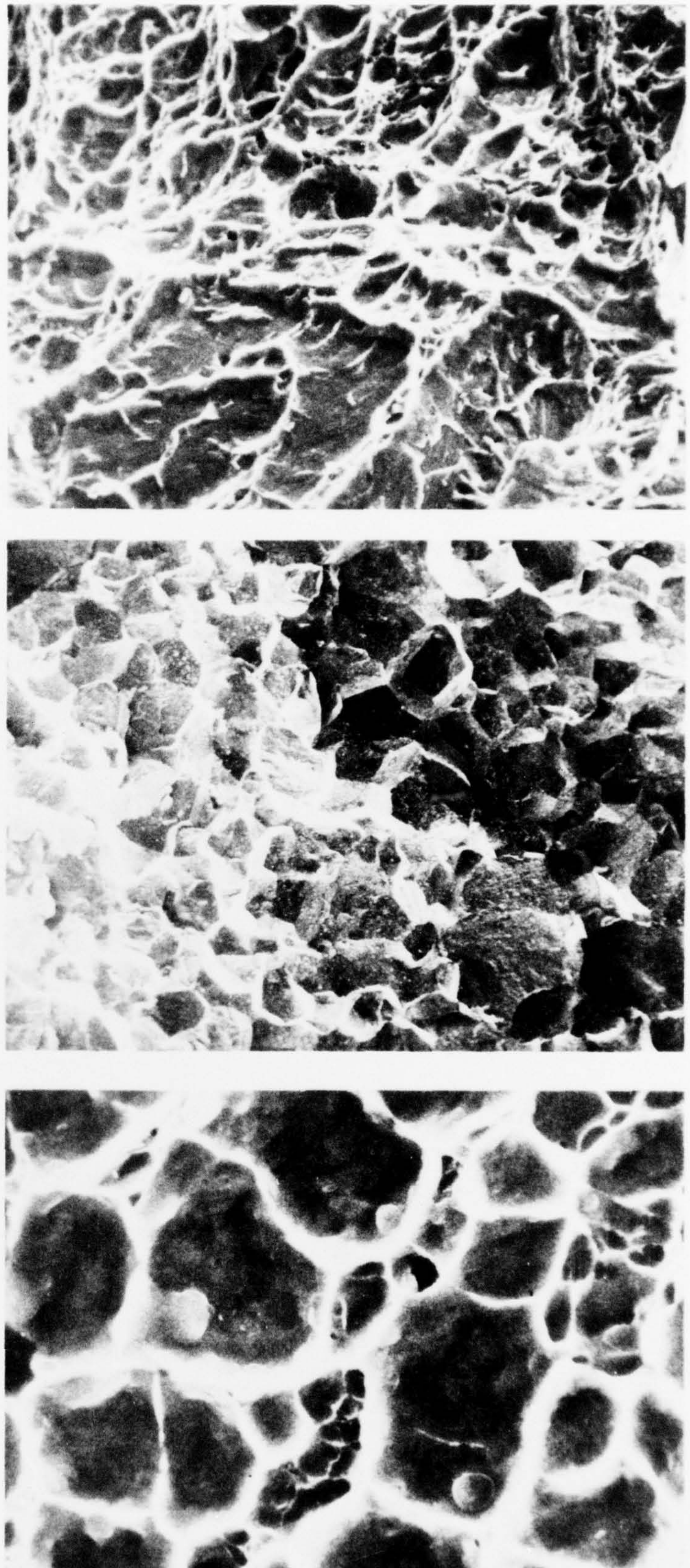


(a)
2000X
$R_c = 55$
$K_{Ic} = 39$
ksi-in$^{1/2}$

(b)
2000X
$R_c = 51$
$K_{Ic} = 65$
ksi-in$^{1/2}$

(c)
500X
$R_c = 44$
$K_{Ic} = 103$
ksi-in$^{1/2}$

Figure 29.

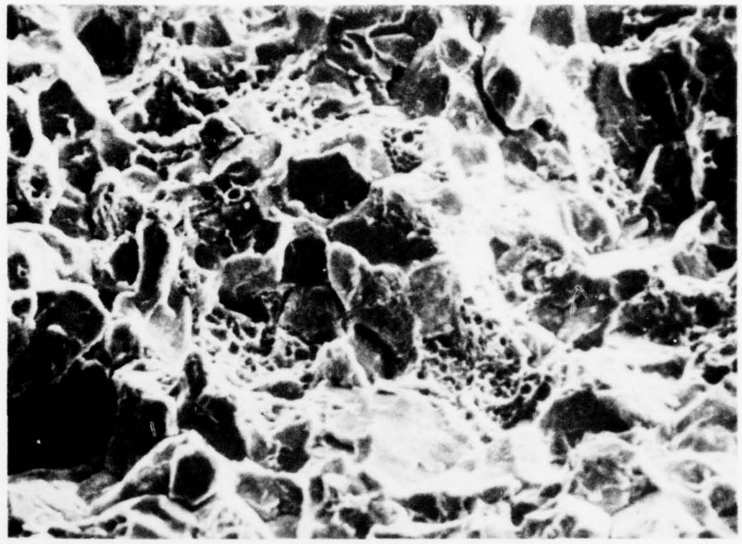


(a)
2000X
$R_c = 52$
$K_{Ic} = 85$ ksi-in$^{1/2}$

(b)
50X
$R_c = 55$
$K_{Ic} = 53$ ksi-in$^{1/2}$

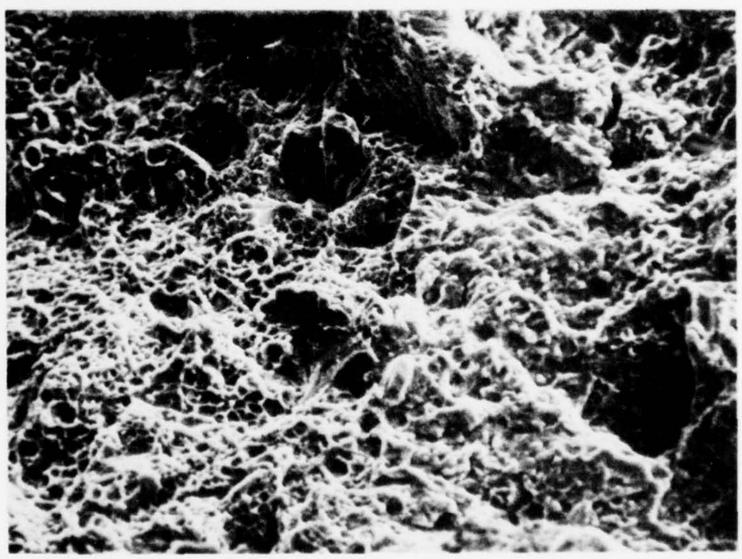
(c)
5000X

Figure 30.



(a) 1000X

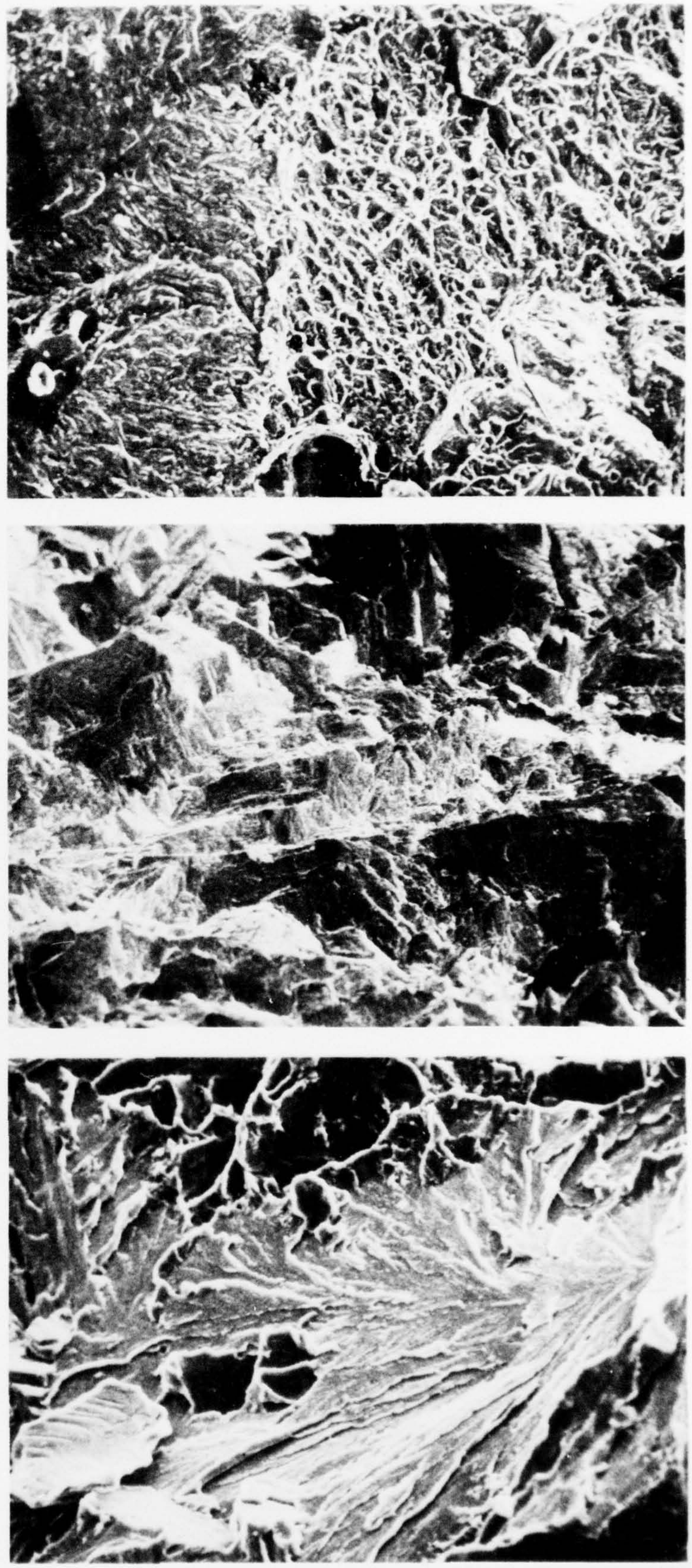
$R_c = 55$
$K_{Ic} = 40$
ksi-in$^{1/2}$



(b) 1000X

$R_c = 47$
$K_{Ic} = 52$
ksi-in$^{1/2}$

Figure 31.



(a)
500
R_c
K_{Ic}
ksi

(b)
100
R_c
K_{Ic}
ksi-

(c)
100
R_c
K_{Ic}
ksi

Figure 32.

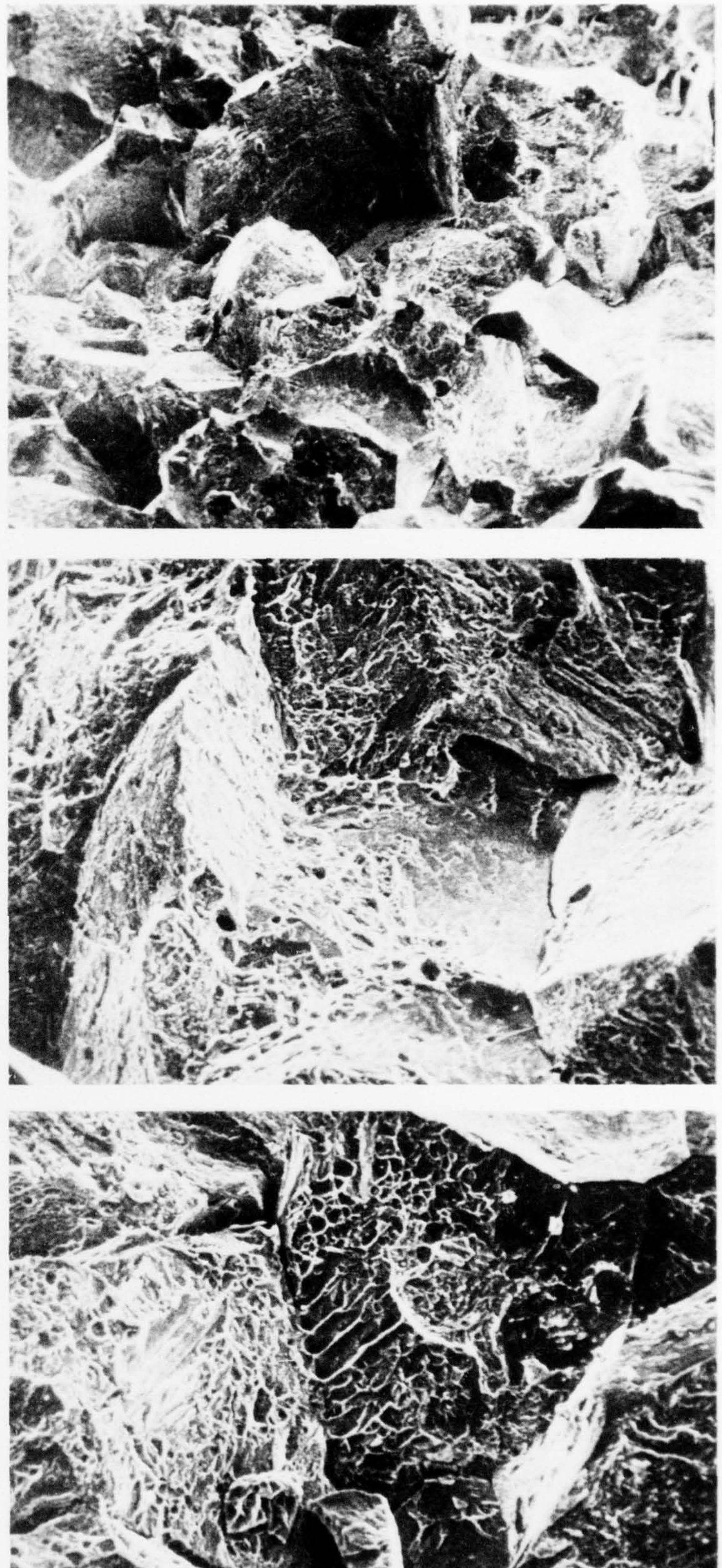
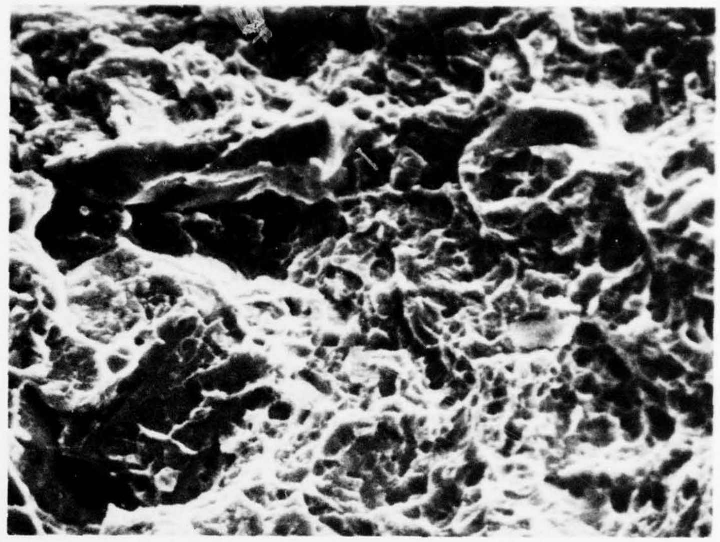
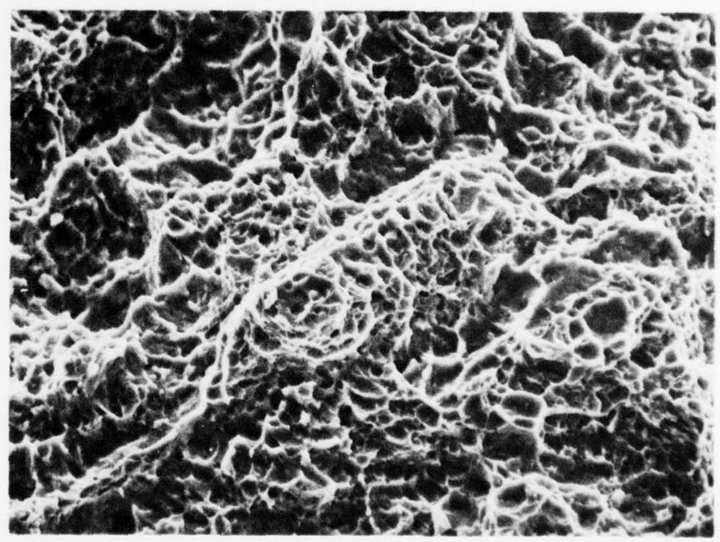


Figure 33.



(a) 2000X

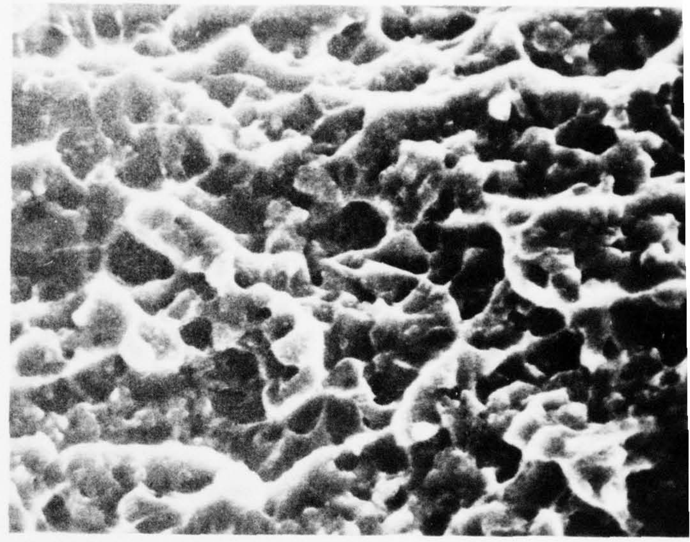
$R_c = 56$
$K_{Ic} = 37$
ksi-in$^{1/2}$



(b) 2000X

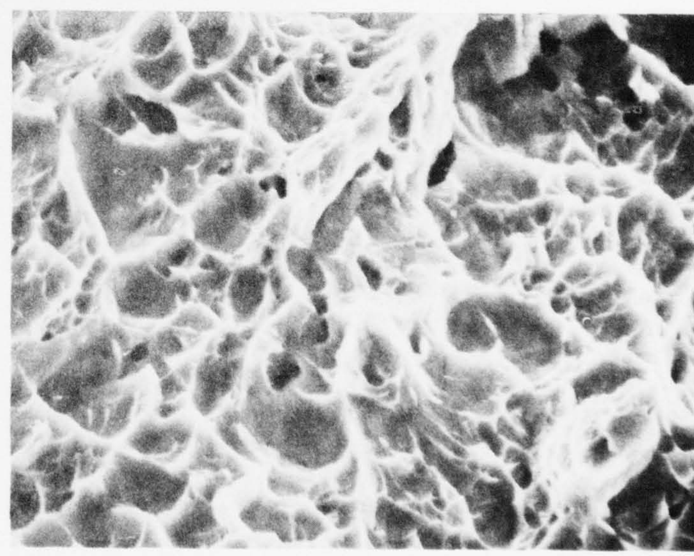
$R_c = 52$
$K_{Ic} = 68$
ksi-in$^{1/2}$

Figure 34.



(a) 5000X

$R_c = 57$
$K_{Ic} = 44$
ksi-in$^{1/2}$



(b) 5000X

$R_c = 54$
$K_{Ic} = 86$
ksi-in$^{1/2}$

Figure 35.

END

DATE

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